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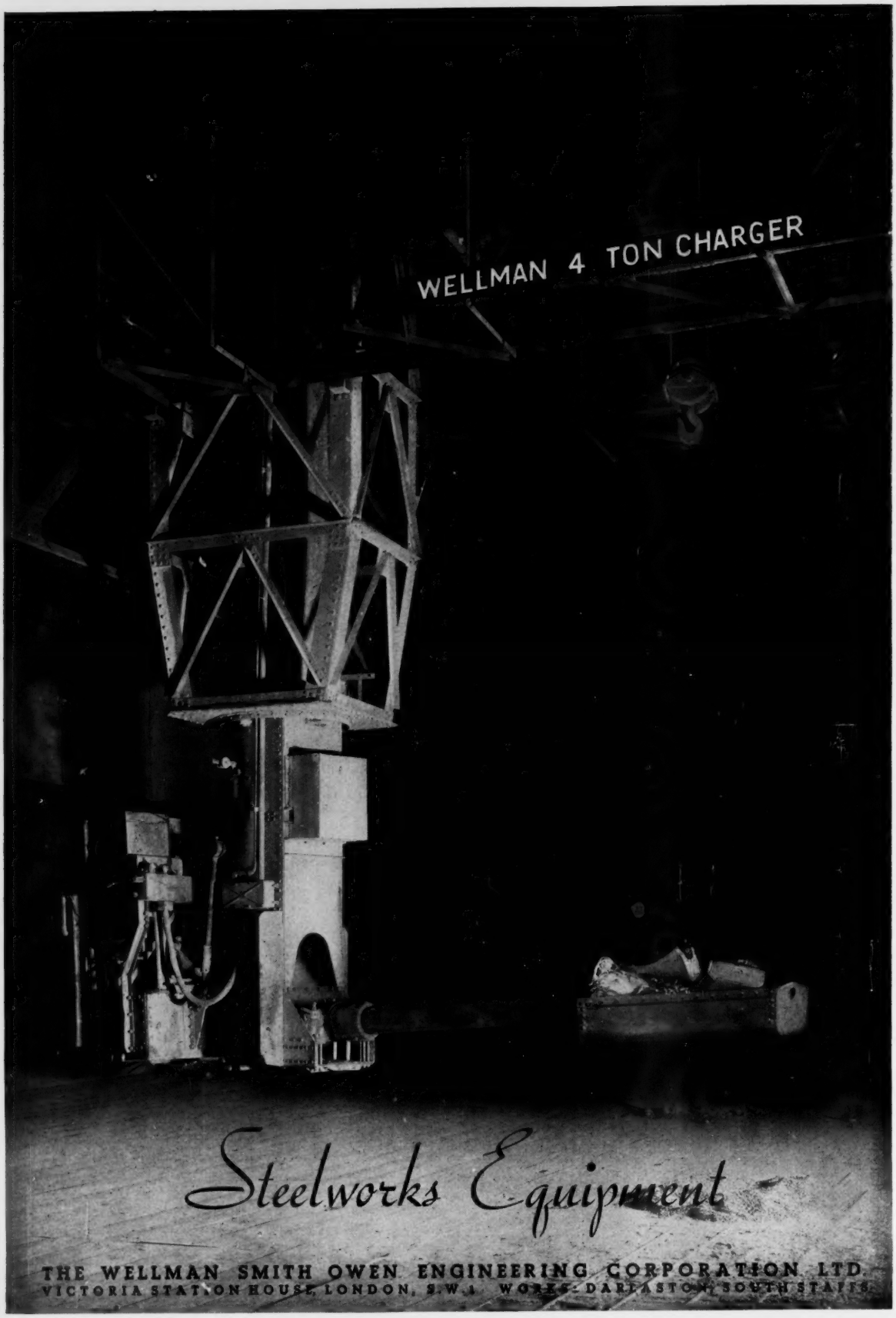
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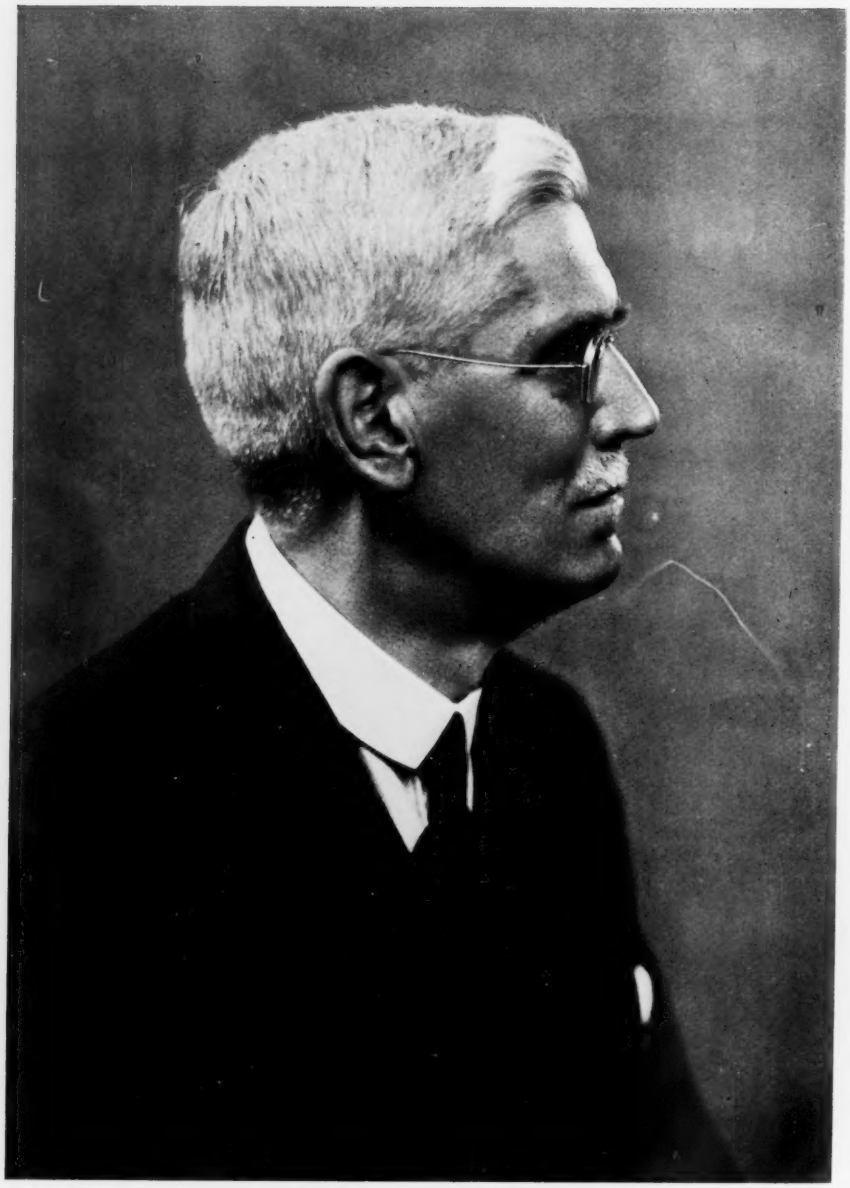
SIR WILLIAM LARKE, K.B.E., D.Sc.

SIR WILLIAM LARKE, K.B.E., D.Sc., received his early engineering training with H. F. Joel & Company and Siemens Bros. & Co., Ltd., Woolwich. In 1898, he joined the British Thomson-Houston Company, and acted as Engineer and Manager of their Power and Mining Department from 1899 to 1912; in 1912, he was appointed Executive Engineer and served in that capacity until 1915. He then joined the Ministry of Munitions and occupied various responsible positions before being appointed Director-General of Raw Materials. He received the O.B.E. in 1917, became C.B.E. in 1920 and was knighted in 1921.

In 1922 he was invited to become the Director of the National Federation of Iron and Steel Manufacturers, which was formed as part of an organised effort by British industry to lift the country out of the depression ruling at that time. It soon became evident that some control of the import position was necessary if industry were to regain its prosperity, and this involved the Federation and its Director in a campaign, designed to impress upon the country and the Government of the day, the soundness of these views. The prosecution of the campaign involved the creation of an organisation competent to produce the economic and other information on which the Federation policy was based. Co-operation within the industry was fostered, in order that its economic interests be protected and its technological efficiency increased. The organisation of co-operative research was undertaken by the Federation in 1924 and the activities of those earlier days have resulted in the present British Iron and Steel Federation and the British Iron and Steel Research Association. The latter carries on the work of the Iron and Steel Industrial Research Council, of which Sir William was Chairman from 1938 to 1945.

Sir William became a member of the Iron and Steel Institute in 1904 and has served for many years as a member of Council and on various Committees of the Council, notably the Finance and General Purposes, Publications and Joint Library Advisory Committees. He was elected Vice-President in 1934, becoming Honorary Vice-President in 1946. He has also been President of the Institute of Fuel, of which he was one of the founders, of the Junior Institution of Engineers, of the Electrical Research Association and of the Institute of Welding. He played a prominent part in the formation of the Welding Research Council in 1935, of which he was Chairman until the formation of the British Welding Research Association, of which he was elected the first President in 1945. He has also been Chairman of the Scientific Instrument Research Association and is at present Chairman of the British Standards Institution and of the Industrial Research Committee of the Federation of British Industries.

His many services to industry and unfailing appreciation of the value of research were recognised by Durham University in the conferment of the honorary degree of Doctor of Science. A further tribute to his services to science and industry is now paid by the Council of the Iron and Steel Institute, who have announced the award to Sir William of the Bessemer Gold Medal for 1947.



Vandyk

Sir William J. Larke, K.B.E., D.Sc.
Bessemer Gold Medallist, 1947

IRON AND STEEL
INSTITUTE PAPERS

PLANT DESIGN · MANUFACTURE · RESEARCH · DEVELOPMENT

Graphite Formation in Cast Irons and in Nickel–Carbon and Cobalt–Carbon Alloys*

By H. Morrogh† and W. J. Williams, B.Sc.†

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* Received 4th May, 1946. † British Cast Iron Research Association, Birmingham.

Synopsis

An attempt has been made to give a concise statement of as many as possible of the features of graphite formation in cast irons which need explanation, together with a critical examination of the numerous theories which have been developed from time to time. A detailed description of the metallography of graphite in relation to its crystal structure is given, and the solidification of flake-graphite-containing irons is described.

The process of undercooling in cast irons is examined with reference to a number of problematical points and several new phenomena with respect to graphite formation in cast irons have been introduced. The effects of titanium and zirconium in refining the graphite depend upon the presence of the carbides of these elements. Two types of inverse chill have been demonstrated.

The process of graphite formation in nickel–carbon, nickel–iron–carbon, and cobalt–carbon alloys has been studied, and the analogies between the mechanism of the process in these alloys with that in cast irons have been worked out. Undercooling in all these alloys and in cast irons takes place by the solidification of a eutectic of solid solution + carbide, and the carbide phase decomposes after solidification. Coarse graphite structures solidify as a eutectic of solid solution + graphite. Nickel–carbon and cobalt–carbon alloys tend to give coarse graphite when they contain more than a certain amount of sulphur. In the absence of sulphur additions, spherulitic nodular graphite structures may be produced in nickel–carbon, nickel-rich nickel–iron–carbon, and cobalt–carbon alloys by the addition of calcium and/or magnesium in a wide variety of forms. The manner in which these spherulites form has been studied.

The formation of spherulitic nodular graphite in cast irons without heat-treatment has been shown to occur. The conditions necessary for this have been outlined, and spherulites have been produced artificially without heat-treatment.

The undercooling of cast irons and of similar alloys has been discussed at length, and the more general aspects of the particular type of undercooling have been emphasized.

Introduction

The presence of graphite in grey cast iron, and to a lesser extent malleable cast iron, confers upon the material its characteristic properties. The relatively low tensile strength and poor shock resistance on the one hand and the good damping and free-machining properties on the other are largely determined in grey cast iron by the presence of graphite. By modifying the amount and shape of the graphite the properties of the material can be considerably changed. The mode of occurrence of graphite in various cast irons and pig irons is known to vary in a manner not always explicable in terms of the known facts or current theory. These variations have very important practical and commercial repercussions. Normally, a grey cast iron has graphite in a lamellar form of some type, but it is well known that if the graphite phase is in the nodular form, as in a heat-treated malleable iron, a material of improved shock resistance and ductility is obtained. In this paper it is the prime concern of the authors to deal with the formation of graphite during the process of solidification and subsequent cooling of castings, and not with the formation of graphite during a heat-treatment process, although, as will subsequently be shown, the two are not disconnected. It is the authors' opinion that there can be produced in cast irons a wide variety of graphite formations which have not yet been reported, still less explored and commercially produced, and which make imperative important changes in the theory of graphite formation.

Until the mechanism of graphite formation in cast iron is fully understood, the full metallurgical potentialities of the material will not be realized. This process presents problems as difficult of solution as any in the whole field of metallurgy. Cast iron is one of the most complex alloys used, and its intrinsic properties are such that experiment is very difficult. There is a very large amount of literature on the subject, but there are still broad gaps in our knowledge. In what follows no attempt will be made to review the literature in any great detail, and reference will be made only to what, in the authors' opinion, is the most recent and most authoritative work. However, an attempt will be made to summarize the existing knowledge of the mechanism of graphite formation and to state the problems and phenomena which need to be explained and linked together. This will be followed by an account of the authors' experimental work; this is by no means complete, and at the present time it appears that the various problems involved require careful experiment which can only be carried out as part of a long-term research programme. The work described here represents the completion of a stage of such a programme.

Much of the experimental work described in this paper deals with the formation of graphite in nickel–carbon and cobalt–carbon alloys. It is

not immediately apparent that this is related to graphite formation in cast iron, but later in the paper reasons are given for the conduct of this investigation and its presentation in this context.

A survey of the literature on the subject reveals a very confusing state of affairs, with numerous theories in existence which in many cases have resulted from work carried out under special experimental conditions, not necessarily generally valid. Attempts to generalize from such theories have involved some contradictions. Before dealing, however, with the debatable issues, the points which are generally accepted as true will be stated. It is significant that such a statement can be devoted only to the broadest issues.

It is well known that cast irons are essentially iron–carbon–silicon alloys containing also varying amounts of manganese, sulphur, and phosphorus as an invariable rule, and that frequently nickel, chromium, copper, molybdenum, titanium, vanadium, arsenic, etc., are present singly and in various combinations. When considering cast irons it is convenient to refer them to the iron–carbon–silicon ternary system, but cast irons differ in many significant ways from alloys of this system. This difference must always be borne in mind. For the purpose of the subject under consideration it can be said that the equilibrium diagrams for the binary iron/iron-carbide and the iron–silicon systems are known with fair accuracy. The ternary iron–carbon–silicon system under equilibrium conditions is much less clearly defined and presents opportunities for considerable further investigation. The issue is, however, complicated by the fact that under practical conditions the iron–carbon–silicon alloys exhibit a metastability anomaly. Under some conditions of composition and cooling, steels and cast irons have major microstructural phases consisting of iron carbide (cementite) and a solid solution of carbon in α or γ iron. In these cases the carbon is said to be in the combined form. On the other hand, according to its composition and thermal treatment, a steel or cast iron may have major microstructural phases consisting of graphite and a solid solution of carbon in α or γ iron. It is generally assumed that iron carbide (cementite) can be considered as a metastable, or thermodynamically unstable, compound of considerable persistence under certain thermal conditions and compositions. This has led to the postulation of the existence of two similar systems, one the metastable system, in which the carbon not in solid solution is present as cementite, and the other the stable system, in which the carbon not in solid solution is present as graphite. In both cases the carbon is assumed to be in the same condition in the liquid state.

In the case of cast irons the situation is further complicated by virtue of the fact that, during cooling, transformations at different stages in the cooling may occur along metastable or upon stable lines. For instance, solidification may proceed in accordance with the stable system, giving, immediately after solidification, a structure of graphite in a matrix of austenite, but this austenite can subsequently transform to pearlite according to the metastable system, giving as a result the familiar pearlitic grey cast iron. Similarly, solidification may occur partly according to the stable system and partly according to the metastable system, giving a mottled iron.

As a general rule, rapid cooling tends to give phases predicted by the metastable system and slow cooling tends to give phases as predicted by the stable system. Alloying elements, *e.g.*, sulphur, chromium, manganese, vanadium, molybdenum, tellurium, etc., may favour the transformations during cooling and solidification to give metastable products. On the other hand, other elements may enable transformations to occur according to the stable system, for instance silicon, nickel, etc. The reason for this behaviour of alloying elements in cast irons is not yet understood. As far as carbide stability is concerned, elements such as cobalt, titanium, and zirconium may have little effect. The proximity of microstructural graphite to cementite produces further instability, the effect becoming more pronounced the finer the graphite particles. Hydrogen gas and hydrogen in the form of water vapour may, in contact with a melt, give increased carbide stability. In a less easily defined manner it is probable that oxygen also influences carbide stability. The tendency for phase changes to occur according to the stable system is greater at higher than at lower temperatures.

In addition to the dual behaviour of cast irons referred to above, a further complication exists. When a cast iron solidifies in the grey condition it may do so in one of two ways, or in a wide range of combinations of these two ways. The graphite which forms at or near to the minimum solidification temperature—the " eutectic " temperature—may occur in a lamellar form commonly known as flake graphite, or in a finely divided form variously referred to in the literature as supercooled, reticular, undercooled, eutectic-like, dendritic, etc., graphite. At this stage it is not proposed to illustrate the differences between these two structures micrographically. The literature contains many illustrations of the differences, and subsequently in this paper the differences will be

dealt with and illustrated in some detail. The mechanical properties and general behaviour of a cast iron are considerably affected by the type of graphite present. In irons which are cast white and subsequently annealed to decompose the carbide, the resulting graphite is generally in a nodular form. In this form the graphite does not exert such a profound embrittling effect as either of the varieties found in normal grey cast irons and referred to above. The production of nodular graphite forms the basis of the method for manufacturing malleable cast iron. It is generally assumed that the nodular form of graphite can be formed only by the heat-treatment of white cast iron or steel. One of the objects of this paper is to indicate that it is possible to obtain nodular graphite structures in cast irons without heat-treatment.

In what follows, the important anomalies relating to carbide stability and graphite formation will first be discussed.

Of the micrographs reproduced in this paper, Figs. 16–19, 23–25, 29, 31–56, 58, 59, 62, 74–78, 82–84, 86, 87, 90, 92, 94–97, 99, 101–104, 106–109, and 113–123 were taken with a 16-mm. achromatic objective, N.A. 0·28, without an eye-piece. A 16-mm. apochromatic objective, N.A. 0·30, was used for Figs. 9–12, 20, 21, 64, 65, 79, and 81. A 12-mm. achromatic objective, N.A. 0·50, was used for Fig. 68. A 4-mm. achromatic objective, N.A. 0·83, was used for Figs. 13–15. A 4-mm. apochromatic objective, N.A. 0·95, was used for Figs. 66, 110, 124, and 125. A 2-mm. oil-immersion achromatic objective, N.A. 1·3, was used for Figs. 57, 60, 61, 63, 67, 69–73, 85, 88, 89, 91, 93, 98, 100, 105, 111, and 112.

Crystal Structure of Graphite and its Metallography

In many ways graphite is a unique phase in ferrous alloys (and in non-ferrous alloys, with the possible rare exceptions of compounds such as calcium silicide ($CaSi_2$) and molybdenum sulphide (MoS), which have similar layered hexagonal structures). It has a crystal structure differing considerably from those of the other phases present, and this fact no doubt is responsible for many of its peculiar properties. Graphite in cast irons has a structure similar to natural and synthetic graphites,[1, 2] which is layered hexagonal (Fig. 1).

When considering the growth of graphite crystallites in cast irons, it is desirable to keep this structure in mind. The carbon atoms lie in a series of parallel sheets, in each of which they are arranged at the corners of a set of regular hexagons. The sheets are so superposed that only one-half of the atoms in each sheet lie vertically above atoms in the sheet below. Alternate layers lie, atom for atom, normally one above the other. The carbon atoms in the layers are very closely spaced compared with the spacing between adjacent planes. This permits easy slip along these planes and accounts for the characteristic mechanical properties of graphite. Recently the presence of another graphite structure in small amounts in both natural and kish graphite has been described.[3, 4, 5] The metallurgical significance of this has not yet been stated.

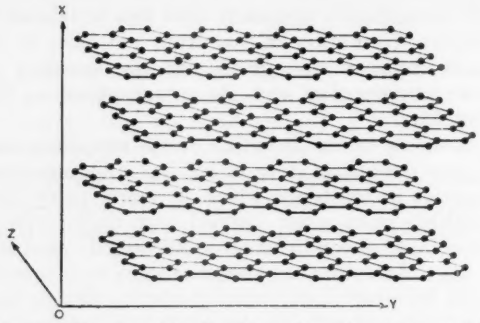


Fig. 1—The graphite lattice. Growth of a graphite flake occurs preferentially in the plane *OZ, OY*

Graphite flakes in cast iron of both the "normal" variety and the "supercooled" variety are essentially basal sections of graphite. With the exception of kish graphite, the flakes are generally distorted in all directions and may be spatially arranged in complex shapes and groupings.[6, 7] In general, however, a graphite flake which appears on a polished microspecimen is so oriented that the basal planes meet the polished surface in a direction approximately parallel to the visual length of the flake if each graphite flake is considered as a single crystal. In other words, the longest dimension of a graphite flake is approximately at right angles to the direction *OX* (Fig. 1).

As the polished surface of a microspecimen of grey cast iron represents a random section through a large number of randomly oriented graphite flakes, there will be a statistical deviation from the orientation relationship given above; hence its approximate character. The probability of a graphite flake occurring with its basal plane parallel to the polished surface exists, but is very slight. When such a case occurs it is impossible to polish the graphite, owing to slip giving exfoliation.

In sections parallel to the basal planes, graphite has a metallic appearance with a high reflectivity and does not exhibit any form of optical anisotropy. Nipper[8, 9] has shown photographs of

basal sections of kish graphite lamellæ presumably scraped from the fractured surface of a sample or from the top surface of a casting head. These clearly reveal the properties referred to above and also show hexagonal markings, as would be expected from the grouping of the atoms in the basal planes. Sections of this type are never seen in polished microspecimens. The sections of graphite flakes normally visible in cast iron are easily polished, have a non-metallic appearance, and exhibit pronounced optical anisotropy and reflex pleochroism.[10] Such polished graphite sections have a light grey-brown colour in unpolarized light, and light grey-brown to black in plane- or partially plane-polarized light, the actual appearance of any graphite crystallite being determined by the reflex pleochroic effect. No phase in cast iron should be termed graphite unless it has been polished and its optical anisotropy and reflex pleochroism confirmed. In the work described in this paper this was carried out as routine in all cases.

So far, it has been assumed that graphite flakes are single crystallites, but in the majority of cases examination at high magnifications under plane-polarized light and between crossed nicols reveals a complex structure which indicates that graphite flakes are composed of a main crystallite with many smaller and differently oriented secondary crystallites in the interior.[1,10,11] It will be shown later in this paper that these internal graphite structures are not peculiar to cast irons, but also occur in nickel–carbon and cobalt–carbon alloys. Hanemann and Schrader[11] have attributed these internal structures in graphite flakes to mechanical distortion during crystallization. The flat graphite crystals are very pliable. The larger crystals of hypereutectic or "kish" graphite freely floating in the melt sometimes form straight lamellæ, whereas the graphite which forms simultaneously with austenite—the "eutectic" graphite—is restrained during solidification and may be distorted by the melt. Mechanical stressing of graphite crystals may give kinking.[12] It is, for instance, well known[13,15] that when long single crystals of hexagonal metal are grown with the glide plane (0001) parallel to the wire axes, the orientation does not favour the usual glide mechanism under tensile or bending loads. When compressive loads are applied the crystals collapse, forming kinks. The arrangement of the kinking in such metal crystals closely resembles the "zigzag" patterns found in a large number of graphite flakes. It has not been possible to prove or disprove the suggestion put forward by Hanemann and Schrader, but several points can be made which are relevant to the issue.

The zigzag formations occur most frequently at bends in graphite flakes, which tends to confirm the mechanical-distortion hypothesis. On the other hand, complex structures not apparently similar to kinking patterns are found throughout some graphite flakes,[10] including perfectly straight "kish" graphite flakes. Again, the zigzag structures are found in portions of graphite flakes not extending through the whole cross-section. For instance, Fig. 9 illustrates an extremely large flake of kish graphite in a blast-furnace-bear sample* viewed under plane-polarized light in its position of minimum pleochroic absorption. The flake will be seen to be substantially free from markings spreading across the whole of its section, but in the regions marked off the zigzag structure is well developed, though on a small scale. This is more clearly revealed in Fig. 10, which shows the same spot between crossed nicols with the graphite flake in its position of maximum brightness. Figs. 11 and 12, which show the same spot between crossed nicols with the graphite flake parallel and at right angles, respectively, to the plane of polarization of the incident light, show the flake to be made up of a number of banded layers. Layers of this type have been shown to occur as the result of the deposition of hypereutectoid and eutectoid graphite on the already existing graphite flakes.[10] However, the zigzag structure appears at the interface between these layers. Figs. 13 and 14 show the area marked off with a full line in Fig. 9 in greater detail at a higher magnification. It seems strange that if these markings are due to mechanical distortion they should not extend across the whole section. Other flakes of kish graphite in the same sample did show markings, each layer having a group of its own, the size of the markings depending upon the width of the banded layer (Fig. 15).

In the case of nodular structures produced by the heat-treatment of white irons, it has been shown[10] that two types exist—the spherulitic and the "graphite-flake-aggregate." The former occurs in irons containing free iron sulphide and the latter in irons containing free manganese sulphide with no iron sulphide. This rule appears to apply for all types of heat-treatment, at temperatures above and below the critical point and with or without pre-quenching. In the course of this paper we shall be largely concerned with the spherulitic type of nodule, and an attempt will be made to show that this structure can be produced in various alloys without any heat-treatment. To the authors' knowledge, the

* Kindly supplied by Dr. J. E. Hurst.

occurrence of the spherulitic structure in cast irons and in the alloys described in this paper is unique in the whole field of metallography, and may perhaps be attributed to the unique crystal structure of graphite. The structure of graphite spherulites is interesting, and typical spherulites are illustrated in Figs. 58 and 63. In common with other spherulitic structures, graphite spherulites each consist of a large number of acicularly shaped crystallites growing outwards from a centre or nucleus. As a general rule each crystallite is so oriented that its basal planes are at right angles to the radii of the spheroid of which each spherulite can be considered to be composed. This orientation relationship is illustrated schematically in Fig. 2 and can easily be confirmed by examination of a nodule under plane-polarized light. This orientation agrees with the orientation rule for spherulites given by Jansen,[85] who showed that in spherulites the most widely spaced crystallographic planes tend to be oriented at right angles to the spherulite radii. Whilst this is generally true, a few crystallites can be detected in most nodules which do not obey the rule. In whiteheart malleable iron it is frequently possible to find iron sulphide particles at the centres of temper-carbon nodules. When the polished section lies above or below the centre of a spherulitic carbon nodule, the crystallites at the centre of the exposed surface are so oriented that their basal planes lie parallel to the polished surface, and so they cannot be polished. This unpolished zone must not be mistaken for a nucleus.

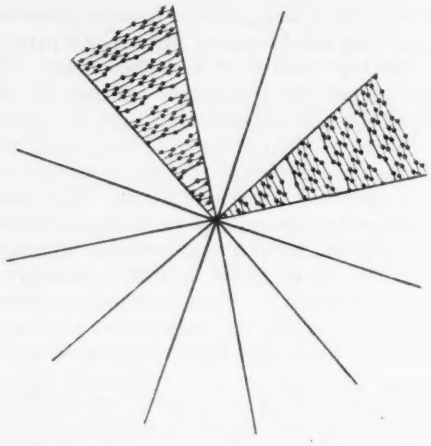


FIG. 2—Schematic representation of the orientation of graphite crystallites in spherulitic carbon nodules

THE SOLIDIFICATION OF CAST IRONS CONTAINING COARSE FLAKE GRAPHITE

Before proceeding further, a description will be given of how, in the authors' opinion, grey cast irons solidify to give flake graphite. The major part of this will perhaps be well known, but many arguable points will necessarily be mentioned. The description has been put in this manner in order to give a defined foundation upon which to erect subsequent arguments. The point to observe is that this description refers solely to irons containing coarse flake graphite, and this will be emphasized again. White irons must necessarily have different treatment, and it is the subsequent purpose of this paper to show that the solidification of undercooled or fine graphite irons is a more complicated process.

The iron–carbon–silicon system shows a trough originating at approximately 1130° C. and 4·3% of carbon in the iron–carbon side of the diagram. With increasing silicon contents this trough moves to higher temperatures and lower carbon contents. In the stable iron–graphite system the point at 1130° C. and 4·3% of carbon represents a eutectic of austenite and graphite. The effect of silicon upon this complex may be represented by the following equation for silicon contents of up to 3% [16]:

$$\text{Carbon, \% in binary complex} = 4{\cdot}23 - \frac{\text{Silicon, \%}}{3{\cdot}2}$$

In iron–carbon–silicon alloys solidifying in the stable system, this binary complex consisting of flake graphite and austenite has an appearance very different from that of most eutectics, and this has caused many workers to suggest that it does not originate at this type of phase change. However, thermal analysis shows that this phase change does take place at approximately a constant temperature and that it possesses many characteristics of a true eutectic. In agreement with the phase rule, most equilibrium diagrams indicate the transformation liquid $\longrightarrow$ graphite + austenite to take place over a range of temperature. Whether this is so is, however, a doubtful point. Equally doubtful is whether the transformation takes place at constant temperature, as the process is generally accompanied by undercooling phenomena which tend to mask the true thermal effects. For all practical purposes it can be considered as an isothermal transformation.

In most discussions[17–19] on the iron–carbon–silicon system, and in almost all attempts to draw it, it is assumed that carbon enters the system as a carbide. For iron–carbon–silicon systems yielding flake graphite as distinct from undercooled graphite, this is most probably incorrect. Boyles [20] has shown that flake graphite forms directly from the melt, and this has been amply confirmed by Eash[21]; the present authors' experience is also in agreement with this. Jass

and Hanemann,[16] in their construction of the ternary system, have assumed the eutectic to be graphite–austenite.

In most of the literature on the subject the binary complex graphite–austenite is referred to and dealt with as a binary eutectic. The justification for the term eutectic is arguable, but it should be remembered that in cast irons, which in a condition of least complexity are Fe–Si–C–P–Mn alloys, the binary complex can theoretically form over a range of temperature and with changing composition. It is perhaps fortunate that these variables appear to be effective to only an inappreciable degree. If this is thoroughly understood it is permissible to refer to the austenite–graphite binary complex as the stable eutectic and, in the metastable system, to the austenite–cementite binary complex as the metastable eutectic. This terminology will be adhered to throughout this paper and is in line with conventional procedure. Alloys containing less carbon than $4{\cdot}23\%$ — (Silicon, %,/$3{\cdot}2$) will be termed hypo-eutectic, and alloys with more carbon than indicated by this formula will be termed hypereutectic. In passing, it should be noted that other elements, notably phosphorus and chromium, also influence the amount of carbon in the eutectic, lowering its value. The effect of phosphorus is fairly well known, and it is generally assumed that the carbon content of the eutectic is lowered by $0{\cdot}3\%$ by 1% of phosphorus.

Hypereutectic irons solidify by the direct formation of graphite from the melt in the form of kish. Owing to its low relative density this phase tends to rise to the surface of the melt, but when entrapped by the metal it generally appears in the microstructure as characteristic long straight flakes. This phase is deposited over a range of temperature, starting at the liquidus surface, until the carbon content of the remaining melt is at the eutectic concentration, when a simultaneous crystallization of graphite and austenite occurs. The form of this eutectic graphite in hypereutectic cast irons has many interesting features which will be dealt with later, but as a general rule it occurs in flake form on a finer scale than the primary kish.

In hypo-eutectic cast irons the first phase to be deposited from the melt is austenite. This austenite, a solid solution of carbon in γ iron, crystallizes from the melt in the form of dendrites. The lengths of these dendrites, for a given wall thickness and composition, vary with the pouring temperature—the higher the pouring temperature the longer the individual dendrite lengths. Under the influence of sharp temperature gradients the dendrites are disposed in a columnar pattern, and with no marked temperature gradients they are equiaxed. The carbon content of this austenite will vary with the silicon content, tending to decrease with increasing silicon. At the eutectic temperature (approximately 1130° C.) and with no silicon, it is generally assumed that austenite can contain a maximum of $1{\cdot}7\%$ of carbon. The solidus surface for the iron–carbon–silicon stable system is not known with any degree of accuracy, and no quantitative statement can be made. The literature contains several suggestions that the solubility of carbon in austenite in equilibrium with graphite is lower than that in austenite in "equilibrium" with iron carbide. Other elements also have an effect on this solubility. With falling temperature, owing to the deposition of austenite, the carbon content of the remaining liquid increases, and at the eutectic concentration the simultaneous deposition of graphite and austenite occurs.

This eutectic forms with a spheroidal crystallization front. The pattern, the number, and the size of these eutectic cells bear no direct relationship to the same characteristics of the primary austenite dendrites. In cast irons containing phosphorus the phosphorus-rich liquid segregates at the boundaries of these eutectic cells and so the resulting phosphide eutectic (steadite) outlines and indicates these cells. The fact that eutectic crystallization proceeds on a spheroidal crystallization front influences the pattern of the resulting graphite, so that the individual graphite flakes tend to be disposed radially in each cell, giving rise to the familiar cluster or whorl formations. This radial disposition is indicative of but a general orientation trend in any given eutectic cell. All the eutectic cells are allotriomorphic, taking shape as a result of contact with neighbouring cells and the availability of space between the primary dendrites. The perfection of the eutectic cells depends to a large extent upon the free space between the austenite dendrites. In all hypo-eutectic grey irons the graphite forms after and hence between the primary dendrites—all graphite in such irons is interdendritic. In low-carbon cast irons the space between the primary dendrites is small, and hence the eutectic cells tend to be very imperfect spheroids and the graphite flakes have to take up positions more markedly dictated by dendrite disposition than by eutectic cell influence. Since in these low-carbon cast irons the primary dendrites determine the graphite-flake pattern, it is obvious that this will be subject to the same influences as the primary dendrites. With high pouring temperatures in low-carbon irons and long dendrites, large series of similarly oriented graphite flakes are obtained.

With low pouring temperatures and short equiaxed dendrites, the graphite flakes tend to be randomly distributed. At this stage it is necessary to recall that only stable alloys yielding flake graphite are being considered. With high carbon contents dendrite disposition is relatively unimportant, and on the macro scale the graphite is again randomly oriented in the form of clusters. In high-carbon irons the dendrite pattern may not be visible in the microstructure, owing to the fact that the primary dendrite austenite is continuous with the eutectic austenite (or their corresponding transformation products exhibit continuity). The same is true of low-carbon irons, but the graphite flakes of necessity outline the dendrites and make their location visible.

We now have a complete statement of the solidification of cast irons yielding flake graphite, according to the stable system. This will provide a defined basis for further discussion. Immediately after solidification, hypo-eutectic and hypereutectic cast irons consist of graphite in a matrix of austenite. Further transformations taking place on cooling will, in practice, depend upon the cooling rate and the silicon content. Two general cases can be mentioned here. If the alloy is cooling slowly, giving stable phases only, or if the silicon content is fairly high (say 3·6%), on cooling below the eutectic temperature carbon is deposited from the austenite in the form of graphite, either upon the already existing graphite or in the form of a Widmanstätten structure. The factors affecting which structure will be obtained have been defined elsewhere [10] (*loc. cit.*, p. 201P). This process will continue until a temperature, the "eutectoid" temperature, is

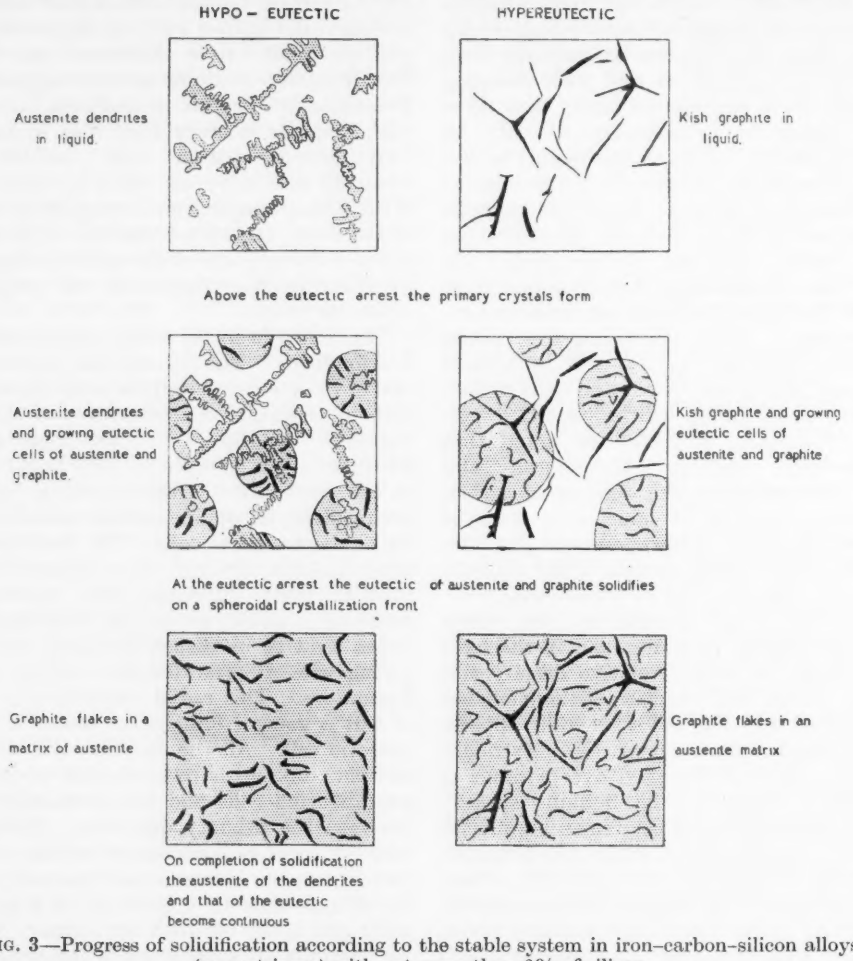

HYPO – EUTECTIC
HYPEREUTECTIC
Austenite dendrites in liquid
Kish graphite in liquid.
Above the eutectic arrest the primary crystals form
Austenite dendrites and growing eutectic cells of austenite and graphite.
Kish graphite and growing eutectic cells of austenite and graphite
At the eutectic arrest the eutectic of austenite and graphite solidifies on a spheroidal crystallization front
Graphite flakes in a matrix of austenite
Graphite flakes in an austenite matrix
On completion of solidification the austenite of the dendrites and that of the eutectic become continuous

FIG. 3—Progress of solidification according to the stable system in iron–carbon–silicon alloys (or cast irons) with not more than 6% of silicon

reached when the austenite simultaneously deposits α iron (ferrite) and graphite over a range of temperature—the critical range. The range of temperature increases and the temperature limits of this critical range rise with increasing silicon. No graphite eutectoid has definitely been seen and this graphite is usually also deposited upon the already existing graphite flakes. When all transformations occur according to the stable system, the microstructure at atmospheric temperature has graphite in a matrix of ferrite. As a general rule, castings—except those with more than about 4·5% of silicon—do not give stable transformation products, as they are not cooled under equilibrium conditions. Frequently, hypereutectoid graphite is deposited on the already existing graphite flakes, but at the eutectoid temperature range the austenite transforms into pearlite (the cementite–ferrite eutectoid), giving a resulting structure of graphite in a matrix of pearlite. At other times the cooling rate and composition are such that the austenite in the proximity of graphite transforms giving stable products, whilst austenite more distant from the graphite flakes transforms giving metastable products. The net effect of this latter condition is to give a structure of graphite flakes in ferrite lakes in a matrix of pearlite.

The process of solidification described so far is illustrated diagrammatically in Fig. 3. It should be remembered that this is intended to apply only to iron–carbon–silicon alloys (or cast irons) containing less than about 6% of silicon. According to the work of Jass and Hanemann,[16] the point of invariant equilibrium occurs at about 6% of silicon and 2·4% of carbon. At this point graphite, γ iron (austenite), δ iron (ferrite), and liquid can coexist in equilibrium. On reaching the eutectic concentration an alloy containing, for instance, 7% of silicon will form the δ-iron/graphite eutectic. This will continue until the point of invariant equilibrium is reached, when the melt will react with the δ solid solution to give γ solid solution. When all the δ has been transformed, the melt will continue to deposit the austenite–graphite eutectic. With about 9% of silicon the same cycle of changes occur, but in this case the reaction between the melt and δ will proceed with the complete elimination of the melt, and so no austenite–graphite eutectic is formed. The deposition of graphite probably accompanies the reaction between liquid and δ, and this graphite tends to occur in fine nodules. With 12% of silicon the δ–graphite eutectic only will form. Fig. 4, taken from the work of Jass and Hanemann and incorporating the suggestions of Scheil,[17] shows an isothermal section of the iron-rich corner of the iron–carbon–silicon diagram at about 10° C. above the transition plane of non-variant equilibrium and illustrates the location of the eutectic trough, the γ + α + liquid and the α+ graphite + liquid three-phase regions, and the point of invariant equilibrium.

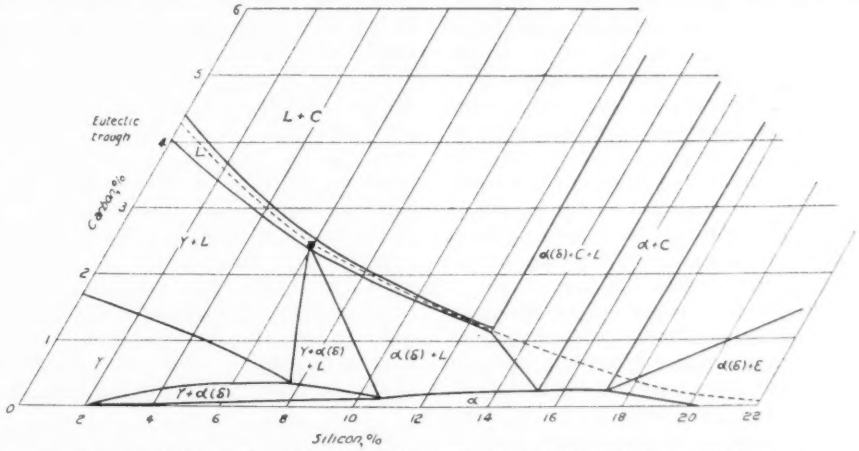


FIG. 4—An isothermal section, just above the plane of invariant equilibrium, of the iron–carbon–silicon diagram. Largely schematic (Jass and Hanemann)

THE ANOMALIES OF GRAPHITE FORMATION

Undercooled Graphite

Thus far, graphite formation in cast iron during solidification has been dealt with solely from the simple point of view of the formation of flake graphite. However, eutectic graphite occurs in another very common form, variously termed supercooled, eutectiform, reticular, undercooled,

etc. Typical examples of graphite of this type in cast irons are clearly shown in Figs. 21, 25, 29, 33, and 64, where the whole of the graphite is in this form. More frequently, in commercial cast irons, flake and undercooled graphite are both formed. The undercooled form shows the greatest tendency to occur in more rapidly cooled parts, that is, in thin sections and at the cast surfaces. Usually when the two forms occur together the undercooled graphite is positioned at the centres of the eutectic cells, with the coarser flake graphite at their extremities (Fig. 16). It would seem that in these cases the undercooled fine graphite derives from the first eutectic to solidify. This undercooled graphite does occasionally, however, occur in sharp juxtaposition to very coarse flake graphite at the central zones of a casting section, that is, in the most slowly cooled portions (Figs. 17, 18, and 19). Frequently in this case the graphite particles of the undercooled graphite at the centre of the section are finer than those near the edge. The mechanism of formation of the undercooled graphite, the factors determining whether this or flake graphite will form, and the reasons for variations in the tendency of irons of apparently the same chemical composition to show varying degrees of carbide stability, are among some of the outstanding problems relevant to the metallurgy of graphite formation in cast iron.

The literature on this subject is extensive, a large amount of experimental work has been accomplished, many observations on routine commercial practice have been recorded, and as a result many theories have been proposed from time to time to explain the phenomena encountered. It will be generally acknowledged that no theory has yet been successful in explaining all the facts. The whole problem has very many facets and, confronted with this, the present authors feel that investigators have tended to devote their attentions to particular aspects. In what follows an attempt will be made to put together as many as possible of the various anomalies which have to be explained, and to correlate the various theories which have been proposed. In addition, various relevant observations made during routine work in the laboratories of the British Cast Iron Research Association will also be included.

Production of Undercooled Graphite by Superheating and by Titanium and CO_2 Treatment

Fine (undercooled) graphite structures have been produced and recorded by a large number of investigators [23-30], and individual reference to all cannot be made. It was observed and confirmed by many that superheating of the melt favoured the formation of undercooled graphite. Norbury [30] showed undercooled graphite irons to have a greater tendency to chill than flake graphite irons of the same composition. The same investigator also pointed out that the eutectic arrest for fine graphite irons took place at lower temperatures than that for flake graphite irons. Norbury and Morgan [31] showed that an iron normally solidifying in a given section to give coarse flake graphite could be made, by an addition of titanium in a suitable form, to yield fine undercooled graphite. It was found that this process could be considerably enhanced by bubbling carbon dioxide through the melt. The same investigators also found that by bubbling hydrogen through the melt, coarse graphite could be obtained when otherwise fine undercooled graphite would be produced. The titanium could be added to the melt in almost any form, but that giving the best yield was ferrosilicon-titanium. Subsequent experience of these investigators in the laboratories of the British Cast Iron Research Association indicated that the presence of titanium in pig irons would explain the apparently inherent tendency of some brands to give fine graphite. For instance, the Scandinavian pig irons Norskalloy and Vantit, both of which contained appreciable quantities of titanium (and vanadium) gave fine graphite structures after many melting operations, including cupola melting. It was also observed that bubbling carbon dioxide through a melt, even in the absence of titanium, will tend to refine the graphite structure.

Forms of Titanium in Cast Iron, and Effect of Zirconium

In irons of normal nitrogen content and containing sufficient manganese to balance the sulphur content, titanium exists in the microstructure as titanium carbide.[32] One of the present investigators (H. M.) has observed that unless the titanium is present as carbide, no refining action is given. When the titanium is fixed as titanium sulphide (owing to insufficient manganese to balance the sulphur), and probably also when completely fixed as titanium cyanonitride, then coarse graphite structures will be obtained. This fact is significant in view of a further finding of the present investigators, *viz.*, that zirconium behaves in a manner similar to that of titanium. In view of the chemical similarity of these two elements, this fact is not surprising. It has been shown [32] that the function of zirconium when added to a normal grey cast iron is to form a probably complex sulphide phase. When the zirconium content exceeds about 0·1–0·2%, the zirconium carbide, ZiC, appears,

and zirconium only seems to exert a positive refining effect when this carbide phase is present. Figs. 20 and 21 illustrate the graphite-refining effect of zirconium. Fig. 20 shows the coarse flake graphite structure in a 3-in. dia. bar, untreated and having the following analysis :

T.C.	Si	Mn	S	P
3·09%	1·58%	0·93%	0·018%	0·05%

Fig. 21 shows the undercooled graphite structure in a 3-in. dia. bar after treatment with ferrosilicon-zirconium to give the analysis :

T.C.	Si	Mn	S	P	Zr
3·13%	3·05%	0·81%	0·011%	0·054%	0·55%

It seems, therefore, to be definitely established that the refining effect of both of these elements is connected with the formation of their carbides.

Von Keil's Silicate Slime, and Norbury and Morgan's Inclusion Theory

Von Keil and his co-workers[33] have attributed the transmission of anomalous characteristics from the melting stock to the casting to the presence of iron silicate in the pig iron, and have suggested that a submicroscopic silicate slime nucleates the melt to give coarse flake graphite, whilst the absence of these fine particles allows of the formation of fine undercooled graphite. Norbury and Morgan,[31] in order to explain the effect of titanium and carbon dioxide, suggested that this process modified the chemical nature of the " inclusions " (meaning silicates and oxides, these being similar to von Keil's silicate slime) in such a manner that they became liquid, whereas they would normally be solid. In this state they could not nucleate the melt, and so fine graphite was produced. Further to support this argument these investigators cited the refining action of copper in excess of liquid solubility. In this case free molten copper was present during the solidification of the eutectic, and so permitted the formation of fine undercooled graphite by surrounding the solid inclusions.

Following this work the literature became liberally interspersed with allusions to " slag inclusions " and " silicate slimes " without any real attempt being made to define these particles with any precision. The idea received some support in the work of Diepschlag [34], who found that melting in contact with aluminous slags had a graphite-refining effect and melting in contact with siliceous slags a coarsening effect. Norbury and Morgan (*loc. cit.*) were unable to confirm this, but it is undoubtedly the experience of the ironfounding industry that slag composition does influence the chilling characteristics and the graphite size of cast iron. The validity of the silicate-slime theory has yet to be proved or disproved, but it will undoubtedly require considerable modification in order to accommodate the other relevant facts. Considerable difficulty has been experienced in advancing this theory further, owing to the difficulty of establishing the identity of the particles necessary for the theory. Microscopically, cast iron is characterized by the almost complete absence of visible oxides and silicates under normal conditions.[32] It may seem surprising that in this respect cast iron and pig iron should be so very much cleaner than steels, but a study of slag–metal equilibria indicates that this should be so. The immiscibility of slag and metal increases at lower temperatures ; at lower temperatures the probability of a slag being solid is greater. The effect of carbon is to increase the region of immiscibility in slag–metal systems. Cast iron tends to be melted and cast at much lower temperatures than steel and in addition has a much higher carbon content, and so the separation of slag is facilitated. By superheating cast-iron melts to over 1700° C., followed by fairly rapid cooling, it has recently been found possible by the present investigators to retain some small particles of glassy silicates in the microstructure. Chemical analysis undoubtedly indicates the presence of oxygen in cast iron, and considerable interest attaches to the mode of occurrence of this element.

In the course of a very lengthy investigation carried out by Norbury and Morgan at the British Cast Iron Research Association, it became apparent that the production of fine graphite by the titanium/carbon-dioxide process* had important influences on the properties of some irons. In irons of coarse graphite and with a relatively high carbon content, an improvement in mechanical properties could be obtained by the refining treatment. Still more important, it was discovered that the production of fine graphite by this method gave sound castings which would otherwise have been porous.[35] This effect is very pronounced, and no doubt explains the soundness of the castings made in permanent moulds by the Eaton–Erb or Holley process.[36] The very rapid cooling of these castings invariably gives fine undercooled graphite. Whilst irons refined by the titanium/carbon-dioxide process are relatively very free from porosity, they show an enhanced tendency to sink on the top surfaces of the casting. Porosity and sinking occur during the solidification of cast iron, and since these vary in amount and type according to whether the graphite is of the flake or undercooled type it seems logical to expect that the mechanism of solidification is

* British Patent Specification 425,227.

different in the two cases. This will be dealt with in greater detail later.

Inoculation

At the same time that attention was focused on the factors affecting the production of undercooled graphite, considerable progress was being made in the production of high-duty irons by the use of controlled low carbon contents of between 2·7 and 3·3%. This experience indicated that undercooled fine graphite structures were unsatisfactory, and led to the development of processes the aim of which was the elimination of these structures and their replacement by flake graphite. This was achieved by a procedure now referred to as "inoculation," which involved treatment of the melt with a relatively small amount of a substance, generally a graphitizer, immediately before pouring. The characteristic of these inoculants is the fact that, added in this way, they have a graphitizing effect much greater than when included in the charge with the other melting stock. This reduction in chilling tendency is accompanied by the elimination of fine undercooled graphite for normal cooling rates. It is impossible to indicate who originated this process, but the use of calcium silicide[37] (which became the basis of the Meehanite process) and ferrosilicon were among the first methods to be adopted. Others have subsequently been developed, such as the use of S.M.Z.,* silicon carbide, and graphite. Almost without exception the graphitizing inoculants contain either silicon or carbon. A feature of the inoculating process is the fact that the effect wears off and does not persist on remelting. If the metal be held in the molten condition too long before casting, normal chilling and undercooled graphite will result.

Undercooling Theory

It has been demonstrated by many investigators[30, 40–43] that eutectic arrest in irons yielding undercooled graphite takes place at a lower temperature than in irons yielding flake graphite. Cooling curves for irons with mixed structures show a recalescence after the initial arrest. These facts, together with the slag-cloud theory of von Keil and that of Norbury and Morgan, to mention only two of the most influential research schools, caused the gradual evolution of a theory which argued that the fine graphite structures were produced by undercooling of the melt in the absence of sufficient or appropriate nuclei.

It is appropriate to consider here the background on which the undercooling theory was evolved. It is well known that pure liquid metals will readily undercool to temperatures below the normal freezing point until a point is reached when crystallization suddenly begins and the temperature rises. This effect can be minimized, and even avoided, by agitation or by nucleating the melt with solid particles which may or may not be of the same composition as the melt. The effect of undercooling† on the form of a cooling curve is illustrated in Fig. 5. In some organic and inorganic compounds the undercooled state may be preserved indefinitely at atmospheric temperatures to give glassy, isotropic, amorphous "solids." No case is known of this in intermetallic alloys. It has been shown for solutions that the region below the normal solubility curve may be divided into two zones. In the region immediately below this curve crystallization will not occur spontaneously, but requires nucleation of the melt. At a certain fixed temperature below the true solubility curve, a zone is reached where crystallization begins without nucleation. For mixtures it is possible to draw a curve, delineating the zone of spontaneous crystallization, which roughly lies parallel to the normal solubility

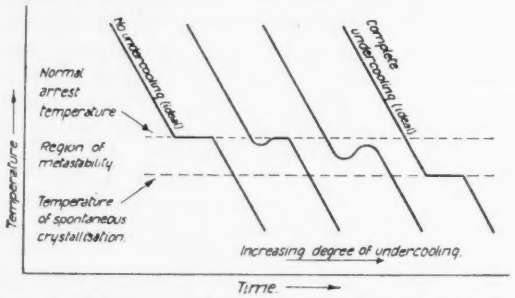


FIG. 5—Effect of undercooling on the form of the cooling curve

curve and which is termed the supersolubility curve. Our knowledge of undercooling in simple eutectiferous systems is limited to the results of Miers and Isaac[44, 45] for the two organic substances betol and salol, and Desch[46] has suggested

* S.M.Z. is a proprietary silicon–manganese–zirconium alloy[39] containing approximately 63% of silicon, 6% of manganese, 6% of zirconium, and 20% of iron. The precise function of the deoxidizing elements is obscure, but this alloy gives better results than ferrosilicon used in amounts to give similar yields of silicon. Significantly however, the performance of ferrosilicon can be improved by mixing with aluminium turnings,[39] and one of the present authors finds this even better than S.M.Z.

† Frequently the term "supercooling" is used synonymously with "undercooling," but the latter is preferred by the present authors as it is felt that "supercooling" carries an implicit contradiction. It is realized that many preferring "supercooling" relate it to supersaturation, supersolubility, and superheating.

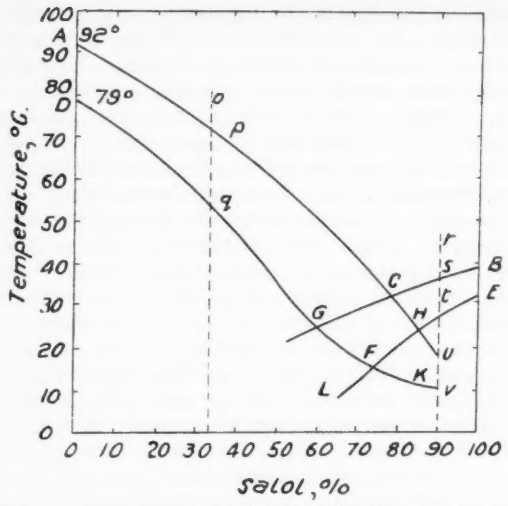


FIG. 6—The system betol–salol, showing the supersolubility curve (Desch)

that these conditions may also apply to metallic alloys. The diagram giving the normal and supersolubility curve for betol and salol (Fig. 6) is reproduced from Desch. The normal eutectic point is at *C*. The two supersolubility curves *DF* and *EF* intersect at a point *F*, termed the hypertectic point. The eutectic and hypertectic need not have the same composition. The area *ACBEFD* is the metastable region requiring nucleation to initiate crystallization. Below *DFE* is the region of spontaneous or labile crystallization. The following quotation from Desch indicates the several possible ways which solidification may proceed owing to undercooling in such a system :

"A mixture represented by the point *o* is cooled, the access of solid matter being prevented. When the point *p* is reached, the solid phase being absent, crystallization does not take place, and the mixture remains liquid until the temperature has fallen to *q*. The supersolubility curve of betol, or more generally of the substance *M*, has now been cut, and crystallization takes place spontaneously if the mixture is shaken. The composition of the part remaining liquid is now represented by a point travelling down *DF*, until *F*, the hypertectic point, is reached. This being the point of intersection with the second supersolubility curve, the second solid, *N*, begins to separate and the whole of the mother-liquor solidifies at the constant temperature of the point *F*.

Taking the same mixture *o*, we may now assume that on reaching the temperature *p*, a crystal of *M* is introduced. This crystal grows, and the point representing the composition of the liquid moves along *pC*. When the eutectic point *C* is reached, however, the eutectic mixture does not solidify as a whole, as it is metastable as regards the substance *N*, the separation of *M* therefore continues in the direction *CH*. The temperature then falls until the curve *EF* is cut, crystallization of *N* begins, and continues until the hypertectic point is reached. The curve *BC* may similarly be prolonged into the metastable region, as shown by *CG*.

In the absence of mechanical disturbance, mixtures may even be cooled below the temperatures indicated by the curves *DF* and *EF* without crystallizing. Thus the mixture *o* may be cooled, under favourable conditions, below *q*, but crystallization is sure to begin before the undercooling has proceeded very far into the labile region, unless prevented by increasing viscosity. The heat developed causes the temperature to rise to *q*, and freezing then proceeds. Under similar conditions, however, the phase *N* may also fail to appear when *F* is reached, so that the labile prolongation *FK* may be experimentally realized. The curve *EF* may be similarly prolonged as at *FL*.

The formation of crystals on the upper curve takes place with comparative slowness, and the temperature may therefore fall low enough for the supersolubility curve to be intersected before the separation of solid is complete. In that case a further crystallization of the same solid phase, but now proceeding rapidly from a large number of centres, may take place. Miers and Isaac found it easy to distinguish by inspection alone the crystals formed in the metastable region from the shower deposited on reaching the supersolubility curve.

It is clear from this description that the eutectic point *C* may entirely fail to appear on the cooling curve of a mixture. In one case, that of a mixture containing 90% of salol and 10% of betol, Miers and Isaac were even able to observe four freezing points, none of which was the eutectic point. Representing such a mixture by *r*, the upper salol curve is crossed at *s*, at which point salol crystals may be obtained by inoculation. Crystals of betol are obtained at *u* by inoculation with that compound, whilst labile showers of the two components may be obtained spontaneously by cooling to *t* and *v* respectively."

It should be noted here that these are not all theoretical possibilities and, as will be shown later, they may be pertinent to the solidification of cast irons.

Possible Modes of Undercooling in Cast Irons

It will be seen that the fine graphite structures obtained by superheating, by rapid cooling, or by the titanium/carbon-dioxide process could perhaps be explained by undercooling according to the pattern outlined above. In the absence of nuclei it may be assumed that hypo-eutectic cast irons could undercool to the supersolubility curve, austenite being deposited down the supersolubility curve until the hypertectic point is reached and its corresponding binary complex is formed (Fig. 7*a*). If austenite formed along the normal liquidus and continued along its extension until the supersolubility curve on the hypereutectic side was reached, another phase would have to be deposited before the hypertectic point was reached (Fig. 7*b*). The present authors

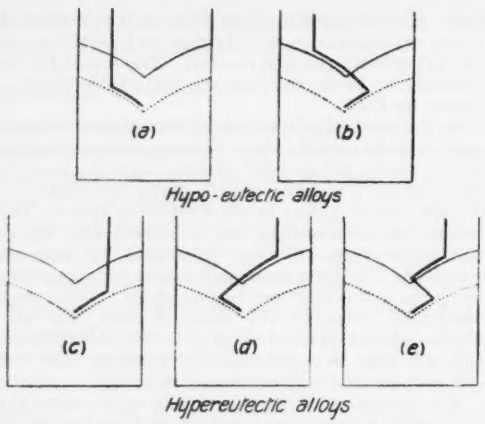


FIG. 7—Possible ways by which undercooling could occur in cast irons

have no definite knowledge of structures indicating that this has occurred, although when taking cooling curves of 50-g. samples of iron giving fine graphite, a change in slope was observed in some cases on the portion of the curve immediately before the eutectic arrest. A typical case is shown in Fig. 8. This change of slope begins at approximately the temperature of the normal eutectic, and may indicate the deposition of supersaturated austenite and/or a phase from the hypereutectic portion of the supersolubility curve.

Three similar cases of undercooling could apply to hypereutectic alloys (Fig. 7, *c*, *d*, and *e*). Diagrams *a* and *b*, for hypo-eutectic alloys, are essentially similar to diagrams *c* and *d*, for hypereutectic alloys. A case, similar to diagram *e*, can be constructed for hypo-eutectic alloys. Some metallographic evidence is available that the type may occur involving undercooling of hypereutectic alloys along the extension of the line indicating the solubility of graphite in the melt until the supersolubility curve for austenite is reached, and then down the supersolubility curve to the hypertectic point (Fig. 7*c*). This type of undercooling would involve primary phases being deposited characteristic of both sides of the eutectic. Some hypereutectic cast irons do show kish graphite and transformed primary austenite dendrites. Nipper[8, 9] has shown some excellent micrographs illustrating this effect with more emphasis than in any sample prepared by the present authors, and Fig. 22, which is a reproduction of one of Nipper's micrographs, illustrates the point. A large flake of primary kish graphite is seen crossing the field, and this flake is completely insulated from the undercooled graphite by a layer of graphite-free metal which in places has the appearance of a primary dendrite. In some cases completely separated kish and primary dendrites can be seen. Nipper's melts were carried out in graphite crucibles in a high vacuum (below 0·0001 mm. pressure) at about 1640° C., followed by slow cooling. Superheating and vacuum heating are known to favour undercooling, and it would seem that the deposition of metal (austenite) around the kish graphite flakes had prevented them from nucleating the melt. This is not always so, as the following example prepared by the authors will show. A 50-gm. sample of hematite pig iron was melted in a graphite crucible *in vacuo* (approximately 0·00001 mm. Hg), using a high-frequency induction-heating unit. The cooling rate was more rapid than that quoted by Nipper. At the bottom of the crucible and along the vertical walls, well-formed flakes of kish graphite—apparently radiating from the crucible walls—had inoculated the melt, and no undercooled graphite was present (Fig. 23). In the upper part of the melt a segregation of broken kish graphite flakes occurred (Fig. 24), owing to flotation from the lower and central zones. The graphite had presumably been mutilated by turbulence of the melt during flotation. Again, no undercooled graphite was present. The central zones from which the kish graphite had been removed by flotation showed considerable areas of undercooled graphite with well-defined transformed primary austenite dendrites (Fig. 25). Except for one feature (which will be dealt with later), this structure appears typically hypo-eutectic. This melt was saturated with carbon at a high temperature, and, except on the basis of undercooling, it is difficult to see how one portion of the structure could appear hypo-eutectic.

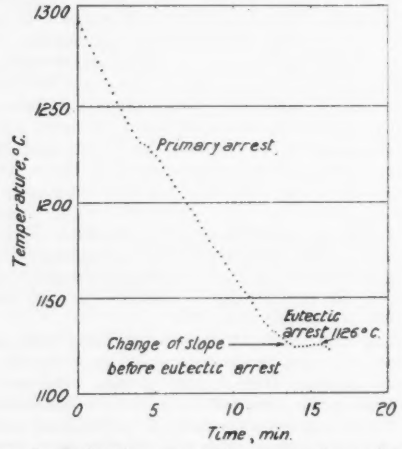


FIG. 8—Inflection on cooling curve immediately before the eutectic arrest

The deposition of a layer of austenite around the primary hypereutectic graphite can be explained by a process other than that referred to above. It is well known that when large and small crystals are in contact with a melt there is a tendency for the larger to continue to grow at the expense of the smaller. This is particularly so when we have primary dendrites surrounded by a solidifying eutectic, the dispersed phase of which is the same as the dendrites. In these cases, in the immediate vicinity of the primary dendrites, the corresponding phase of the eutectic is deposited upon the already existing dendrite, causing it to be surrounded by an unbroken layer of the continuous phase and therefore apparently separated from the eutectic. Desch[48] has illustrated this feature in copper–phosphorus alloys, but iron–carbon alloys are especially interesting in this respect, as is indicated in what follows. In the white irons the dispersed phase of the eutectic is austenite and the continuous phase is cementite, so that white irons of hypo-eutectic composition show this phenomenon. In grey irons the dispersed phase of the eutectic is graphite and the continuous phase is austenite, so that only hypereutectic grey irons would be expected to show this effect. At the eutectoid temperature in white irons the dispersed phase of the eutectoid is cementite and the continuous phase is ferrite, and, as a result, the eutectoid cementite frequently crystallizes on to the already existing eutectic cementite, giving the apparent anomaly of ferrite adjacent to cementite.

From what has been said so far on this point it will be apparent that it is not clear whether undercooling causes the formation of austenite around the graphite, or whether austenite forming round the graphite permits undercooling by insulating it from the melt.

At this stage it should be noted that later in this paper undercooled graphite will be demonstrated to arise from the decomposition of the carbide of a carbide–austenite binary complex shortly after solidification. If the scheme of undercooling outlined above is correct then the hypertectic point is that at which this carbide–austenite binary complex solidifies. Furthermore, the phase deposited down the hypereutectic side of the supersolubility curve must also be a carbide. The unique implications of this will be dealt with more fully in a later section.

Graphite-Nuclei Theory

The silicate-slime theory of von Keil and that of Norbury and Morgan could not explain the inoculating effects of ferrosilicon, calcium silicide, and graphite. Piwowarsky[23] developed a graphite-nuclei theory to explain the increased tendency to undercooling as the result of superheating. This process caused the solution of these graphite particles. This theory was considerably weakened when the same investigator presented evidence[49] to show that graphite dissolves extremely rapidly above the liquidus. Recently, however, the theory has received further support, by Massari and Lindsay,[50] who claim to have demonstrated the presence of traces of residual free carbon in remelted material which decreased in amount with increase in temperature of superheat. Metal from the bath was dribbled directly into cold water and the resulting shot was sectioned for metallographic examination. In support of their claims these investigators produced micrographs which showed black spots, which were termed graphite nuclei. No special tests, such as the determination of the optical properties under polarized light, were carried out to identify this phase as graphite. The present authors emphasize the extreme value of such tests in support of claims which might be put to doubt. Massari and Lindsay also confirmed the powerful inoculating (chill-removing) effects of graphite additions to the melt. This latter point is now well established in commercial practice. Eash[43] has also added his support to the graphite-nuclei theory by suggesting that the inoculating effect of ferrosilicon is due to localized silicon concentrations reducing the carbon solubility. A localized silicon concentration might cause the iron to become hypereutectic and so to deposit kish graphite, which nucleates the melt. It is conceivable that calcium silicide might function in this manner, but it is difficult to maintain that treatment with hydrogen, which also produces flake graphite, would cause local deposition of graphite. This effect is more susceptible to treatment by Norbury and Morgan's theory.

Effect of Gas and Sulphur on Formation of Graphite—The Work of Boyles

In a remarkable series of papers, Boyles[20, 42, 51, 52] demonstrated the mechanism of solidification of hypo-eutectic cast irons. The description already given in this paper is largely in agreement with the findings of this investigator. He demonstrated that hypo-eutectic iron–carbon–silicon alloys and cast irons when melted *in vacuo* tend to give undercooled graphite structures. When remelted in hydrogen this vacuum-melted material was found to give coarse flake graphite, but iron–carbon–silicon alloys behaved somewhat differently from cast irons. This investigator also

indicated that the coarsening effect produced by remelting undercooled graphite alloys in air or nitrogen is due to the absorption of hydrogen, as water vapour, from these gases.

Another very important feature revealed by Boyles was that iron–carbon–silicon alloys which normally give the modified structure behave as cast irons giving coarse flake graphite when an addition of sulphur, or sulphur and manganese, is made. This is a most important finding, and Boyles has suggested that the effect of sulphur is related to its carbide stabilizing action. With sulphur additions beyond a certain percentage it was found that the coarsening action ceased. With reference to the findings of Norbury and Morgan, Boyles suggested that the effect of carbon dioxide in giving undercooled structures was due to its scavenging action in removing hydrogen from the melt. The effect of sulphur additions in causing the formation of large amounts of manganese sulphide in the melt, and so promoting the formation of coarse flake graphite, has been confirmed by the present authors.

That manganese sulphide may nucleate a melt seems highly probable, and the present authors have observed manganese sulphide segregations associated with patches of coarse graphite. Unfortunately, Boyles did not give any precise indications of the types of sulphide phases present in all of his melts. In one series of melts various additions of sulphur were made to base material which contained no manganese. In this case it seems logical to assume iron sulphide to be present. Under these conditions, increasing sulphur gave progressively coarser graphite sizes until a maximum was reached beyond which the graphite became finer. Iron sulphide containing no manganese is liquid during the eutectic solidification, and therefore the coarsening effect cannot be due to solid nucleation. In cast irons with high carbon and silicon contents it is possible that the melt is not completely miscible with liquid iron sulphide, so that some free sulphide may exist. Oxides and silicates, particularly FeO, are known to be soluble in iron sulphide, so that this compound may remove oxygen from the metal by partition.

Eutectic Cell Size and Nucleation

Boyles has also demonstrated that the eutectic cell size is greater and that these cells are much less numerous in undercooled irons than in flake graphite irons. To a considerable extent this fits the facts relevant to the nuclear theory. The number of eutectic cells in a given sample precisely indicated the number of centres at which eutectic crystallization begins. If undercooling in cast iron occurs owing to the absence or deficiency of nuclei, then fewer eutectic cells are to be expected. It is frequently wrongly supposed that nucleation of cast-iron melts solely implies the initiation of the crystallization of graphite. However, in this case we are really concerned with initiating the crystallization of a eutectic. That the graphite phase of the eutectic is the important one for the nucleation is highly probable, and no case has come to the authors' knowledge of austenite initiating the crystallization of the normal flake graphite eutectic. The number of graphite flakes present in a sample is not primarily or directly a function of the state of nucleation of the melt. Nucleation of the melt merely affects the number of centres at which eutectic crystallization begins, and as this determines the size of the eutectic cells it has an indirect effect on graphite size. The radius of a eutectic cell represents the maximum possible length of a eutectic graphite flake, and as the carbon content of the alloy is reduced the realization of this maximum is reduced, owing to a higher probability of an austenite dendrite opposing eutectic crystallization in any one direction. When the nucleation and carbon content of a melt are fixed, the graphite-flake size then becomes a function, within the above limits, of the cooling rate and (an associated factor) turbulence of the solidifying melt. It has frequently been wrongly supposed that the fine undercooled graphite flakes being present in large numbers indicates a high state of nucleation, but the number of eutectic cells indicates the state of nucleation and the uniform fine nature of the graphite is an indication of the spontaneity of its formation. It should be noted that no explanation of the formation of this fine graphite has been given, except in so far as undercooling is a necessary concomitant.

In what has been said so far the term "nucleus," implied by "nucleate," has not been used in any other sense than that indicating a location in the melt where physical and/or chemical conditions are most favourable to crystallization. The value of a particular location for initiating crystallization must depend upon the cooling rate, that is, upon the extent to which the melt is undercooled below the true liquidus. For inoculated irons there is a considerable range of cooling rates within which the nuclei function to initiate flake graphite crystallization, but with very rapid cooling rates low-silicon irons will become white and high-silicon irons will give undercooled graphite. For instance, in permanent-mould irons both kish graphite and late additions of ferrosilicon or

FIG. 9—Graphite flake in blast-furnace bear, in position of minimum pleochroic absorption. Under plane-polarized light. Unetched × 200

FIG. 10—Same spot as in Fig. 9, but between crossed nicols and with graphite in brightest position, *i.e.*, at 45° to plane of polarization of incident light. × 200

FIG. 11—Same spot as shown in Figs. 9 and 10, but with graphite flake parallel to plane of polarization of incident light, *i.e.*, in the same position as for minimum pleochroic absorption. Between crossed nicols × 200

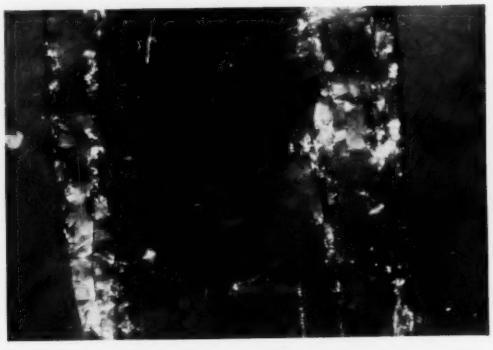


FIG. 12—Same spot as shown in Figs. 9 to 11, but with graphite flake at right angles to the plane of polarization of incident light. Between crossed nicols × 200

FIG. 13—Spot marked off in Fig. 9 under same conditions, but at a higher magnification. × 500

FIG. 14—Spot marked off in Fig. 9 at same magnification as in Fig. 13, but between crossed nicols. × 500

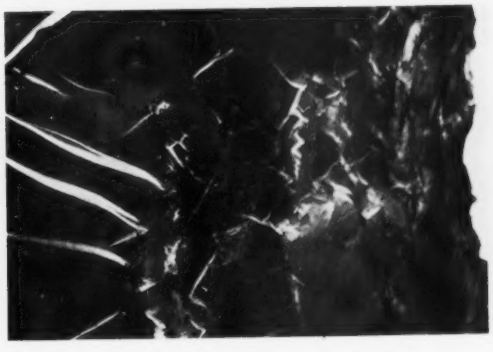


FIG. 15—Flake of kish graphite in blast-furnace bear, with zigzag markings in outer banded layers and at the centre. In position of maximum pleochroic absorption under plane-polarized light. Unetched × 500

(Micrographs reduced to three-quarters linear in reproduction)

[*Murrough and Williams*
[*To face p.* 336

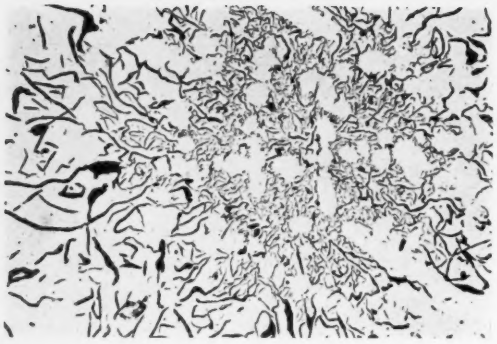


FIG. 16—Cluster-type graphite with undercooled graphite at the centre and normal flake graphite at the periphery of the cluster. Unetched × 60

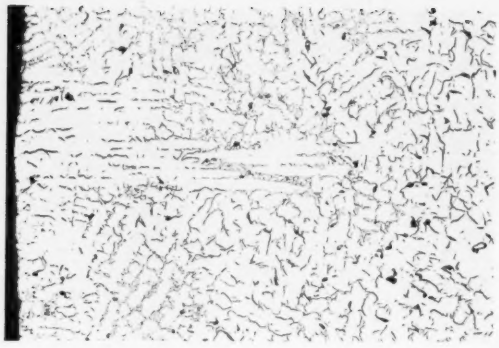


FIG. 17—Graphite structure near the edge of a 1·2-in. test-bar. Fine flake graphite, with traces of undercooled graphite near the edge. Unetched ×60

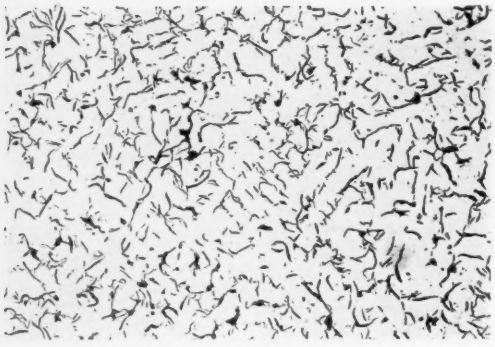


FIG. 18—Same bar as illustrated in Fig. 17, but showing the graphite structure midway between the edge and centre. Medium-sized flake graphite. Unetched × 60

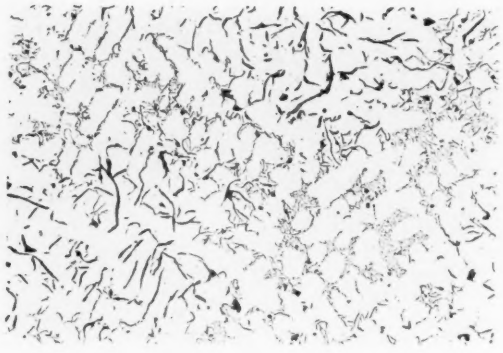


FIG. 19—Same bar as illustrated in Figs. 17 and 18, but showing the graphite structure at the centre of the section. Coarser flake graphite with patches of fine undercooled graphite. Unetched ×60

FIG. 20—General structure of 3-in. dia. bar. Untreated. Etched with picric acid × 100

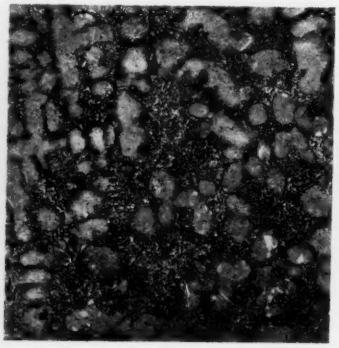


FIG. 21—General structure of 3-in. dia. bar. Same melt as in Fig. 20, but after treatment with ferrosilicon-zirconium. Etched with picric acid × 100

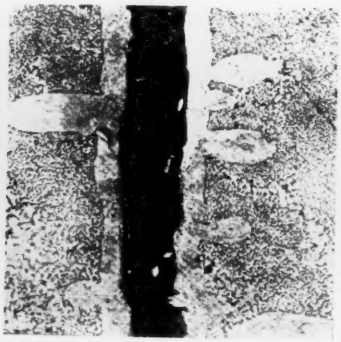


FIG. 22—Primary graphite flake surrounded by transformed austenite and fine eutectic graphite (Nipper). Etched × 200

(Micrographs reduced to three-quarters linear in reproduction)

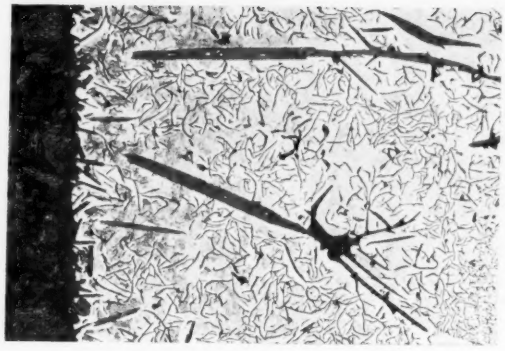


FIG. 23—Well-formed hypereutectic graphite flakes at the portions of ingot near to crucible walls. Ingot was melted in vacuum, using a graphite crucible. Etched with picric acid × 60

FIG. 24—Coarse graphite and fragmented hypereutectic graphite near the top of vacuum-melted ingot. The kish graphite has inoculated the melt in this zone. Etched with picric acid × 60

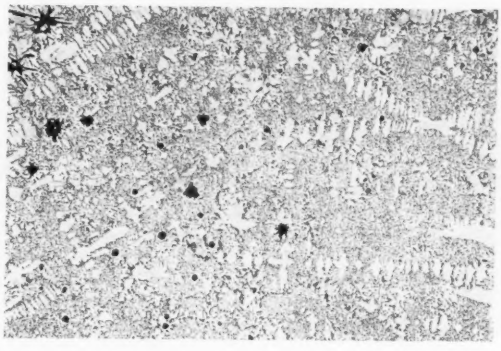


FIG. 25—Central zone of vacuum-melted ingot. Undercooled graphite with transformed primary austenite dendrites. The hypereutectic graphite has floated away from this zone to the top of the ingot. Etched with picric acid × 60

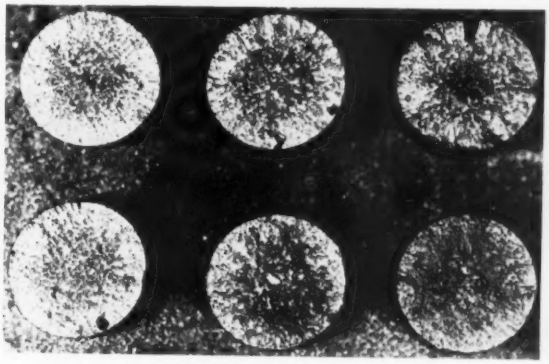


FIG. 26—Bars treated with sulphur, showing effect of pouring temperature on chilling

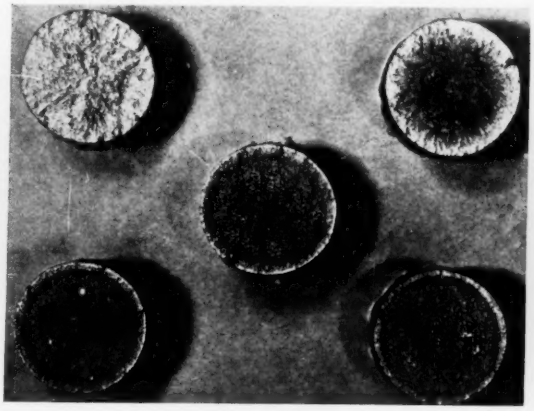


FIG. 27—Effect of pouring temperature upon the chilling of an iron treated with tellurium

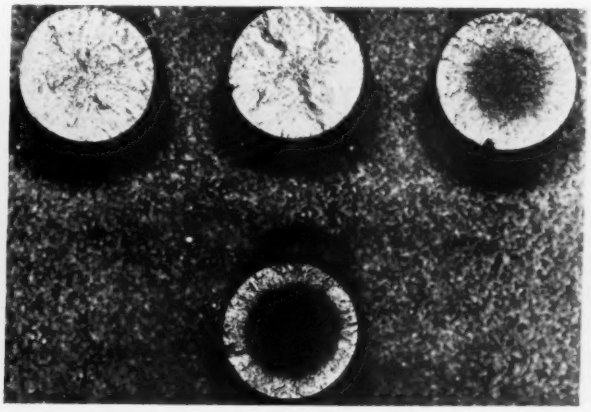


FIG. 28—Effect of pouring temperature upon the chilling of a tellurium-treated iron after remelting

(Figs. 23 to 25 reduced to three-quarters linear in reproduction)

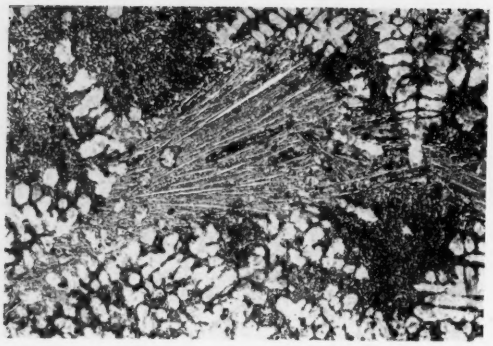


FIG. 29—Inverse chill in an iron with undercooled graphite. Remainder of structure is undercooled graphite in a matrix of pearlite. Etched with picric acid × 60

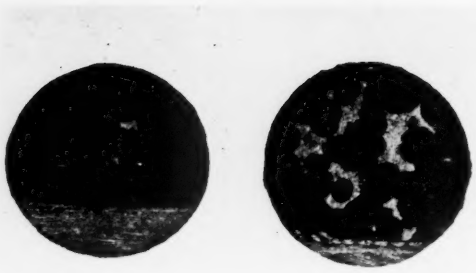


FIG. 30—Fractures of 0·875-in. bars, showing inverse chill

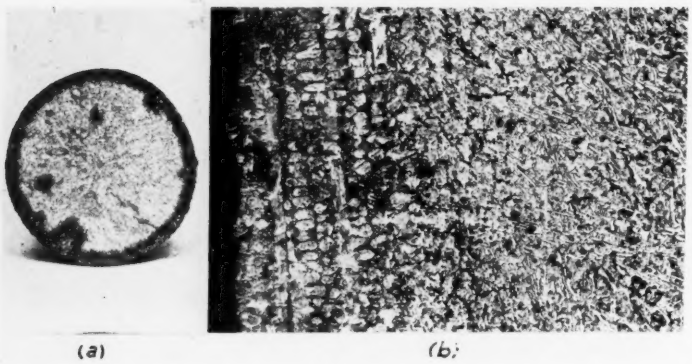


FIG. 31—Influence of sand mould in causing a white iron to have an outer grey rim: (*a*) Fracture of a 0·875-in. dia. test-bar. (*b*) Micrograph of a similar bar showing graphite in the outer rim. Etched with picric acid × 60

FIG. 32—Mixture of flake and undercooled graphite. Unetched × 60

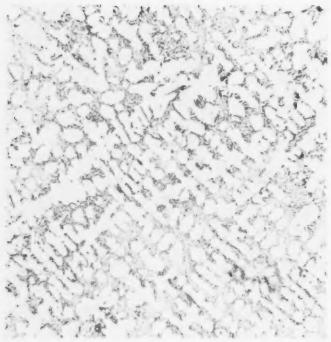


FIG. 33—Same melt as in Fig. 32, but after treatment with 0·2% of bismuth. Entirely very fine undercooled graphite. Unetched × 60

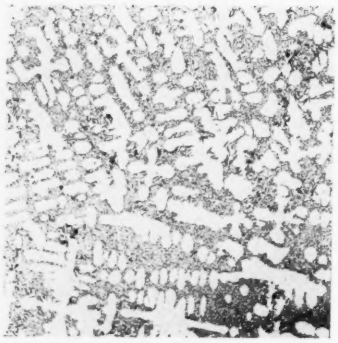


FIG. 34—Hypo-eutectic nickel-carbon alloy. Ingot 1*A*. Melt superheated to 1600° C. Unetched × 60

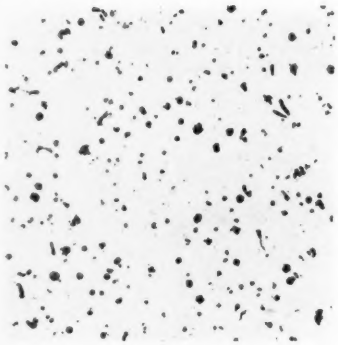


FIG. 35—Hypo-eutectic nickel-carbon alloy. Ingot 1*B*. Same melt as in Fig. 34, but after treatment with calcium silicide. Nodular graphite. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

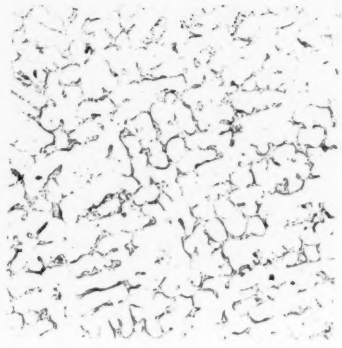


FIG. 36—Hypo-eutectic nickel carbon alloy. Same melt as in Fig. 35, but after treatment with ferrosilicon-zirconium. Unetched × 60

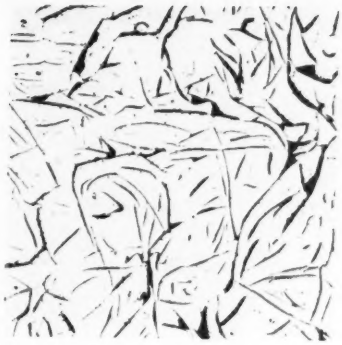


FIG. 37—Hypo-eutectic nickel carbon alloy. Ingot 2*B*. After treatment with tellurium. Coarse flake graphite. Unetched × 60

FIG. 38—Coarse graphite structure in hypo-eutectic nickel carbon alloy after treatment with sulphur. Ingot 4*B*. Small areas of Ni_3S_2 can be seen. Unetched × 60

FIG. 39—Flake graphite structure in ingot 4*C*. Same melt as in Fig. 38, but after addition of 1% of manganese. Unetched × 60

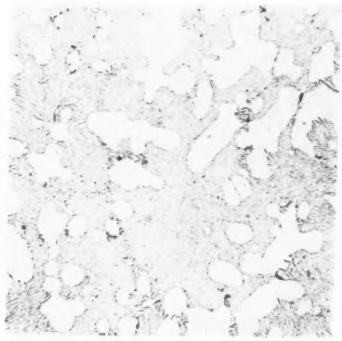


FIG. 40—Hypo-eutectic nickel-carbon alloy. Ingot 5*A*. Superheated to 1600° C. Undercooled-type and coarser graphite. Unetched × 60

FIG. 41—Ingot 5*B*. From same melt as in Fig. 40, but after bubbling hydrogen through the melt. Unetched × 60

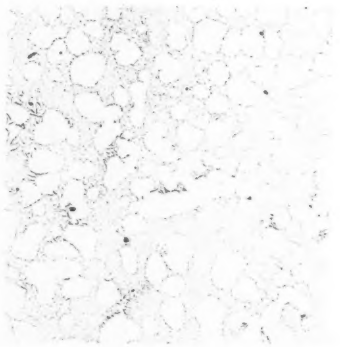


FIG. 42—Undercooled graphite structure in a hypo-eutectic nickel-carbon alloy after treatment with calcium silicide and then with hydrogen. Ingot 6*C*. Unetched × 60

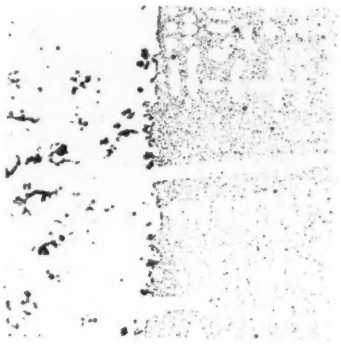


FIG. 43—Undercooled graphite rim at the periphery of ingot 6*B*, treated with calcium silicide. Unetched × 60

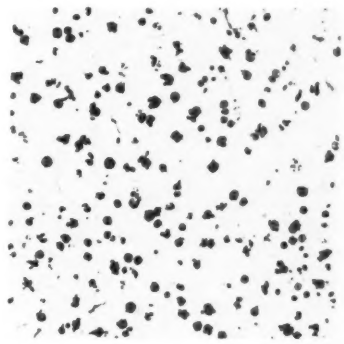


FIG. 44—Hypereutectic nickel carbon alloy. Ingot 8*B*. Same melt as in Fig. 45, but after treatment with calcium silicide. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

FIG. 45—Hypereutectic nickel-carbon alloy. Ingot 8.4. Superheated to 1600° C. Kish graphite flakes in undercooled graphite eutectic with primary solid-solution dendrites. Unetched × 60

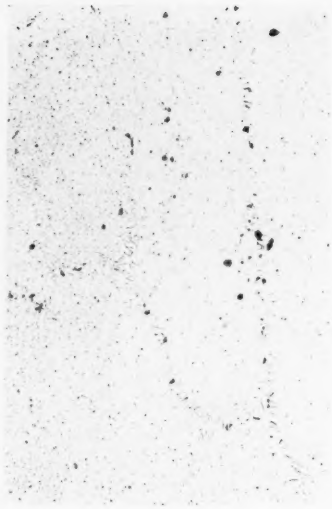


FIG. 46—Wholly eutectic area in ingot 9.4. Hypereutectic graphite has floated to the top of the ingot. Unetched × 60

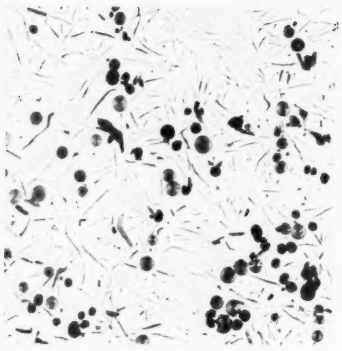
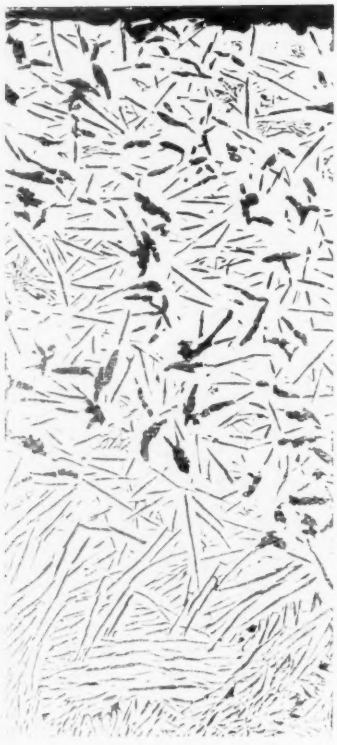


FIG. 47—Area near the top of ingot 9.4; flotation of hypereutectic graphite has occurred. Well-formed spherulites are also present. Unetched × 60

FIG. 48—Hypereutectic nickel-carbon alloy treated with tellurium. Ingot 9*B*. Very coarse flake graphite. Unetched × 60

FIG. 49—Hypo-eutectic nickel-carbon alloy. Ingot 10.4, superheated to 1600° C. Unetched × 60



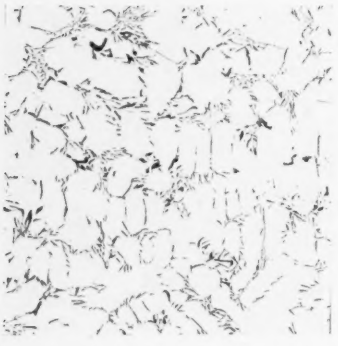


FIG. 50—Hypo-eutectic nickel-carbon alloy. Same as in Fig. 49, but after treatment with titanium and CO_2. Unetched × 60

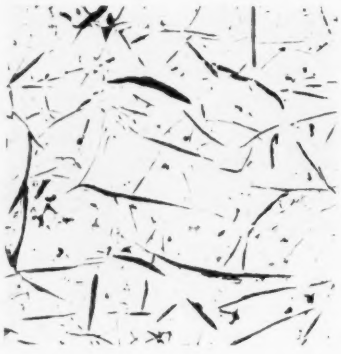


FIG. 51—Ingot made from melt of pure nickel + 2% of C, 2·5% of Si, 1·25% of Mn, 1·3% of P, and 0·3% of S. Coarse flake graphite and lakes of nickel phosphide. Unetched × 60

FIG. 52—Segregation of kish graphite at the top of ingot 16. Note absence of undercooled graphite. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

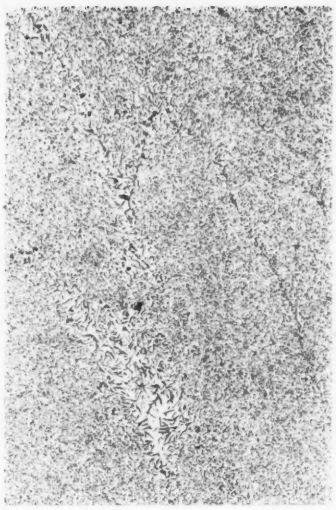


FIG. 53—Central zone in ingot 16. The hypereutectic graphite has floated to the top of the ingot. Unetched × 60

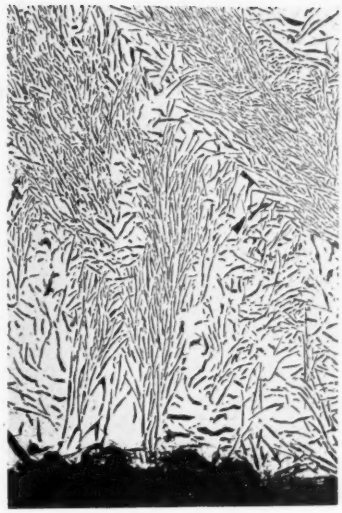


FIG. 54—Coarser graphite near the silica thermocouple sheath in ingot 16. Unetched × 60

FIG. 55—Patch of undercooled graphite in ingot 18. Unetched × 60

FIG. 56—General structure of ingot 18. Nodular graphite in metallic matrix. Etched with neutral ferric chloride × 60

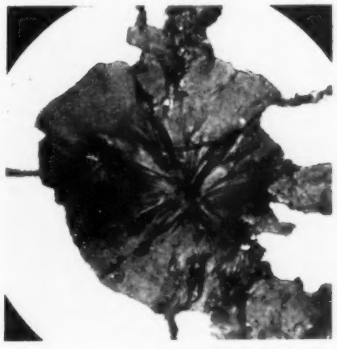


FIG. 57—Spherulitic graphite nodule in ingot 18. Under plane-polarized light. Unetched × 1500

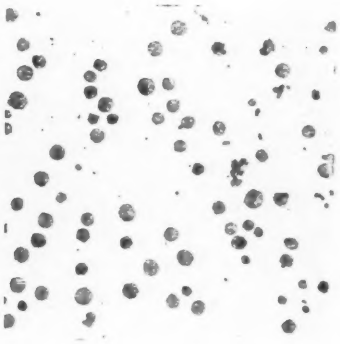


FIG. 58—General structure of ingot 20, treated with calcium silicide. Well-formed graphite spherulites. Unetched × 60

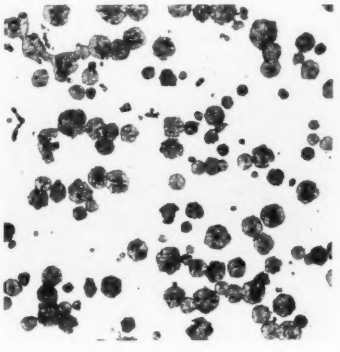


FIG. 59—Segregation of hypereutectic spherulites at the top of ingot 20, treated with calcium silicide. Unetched × 60

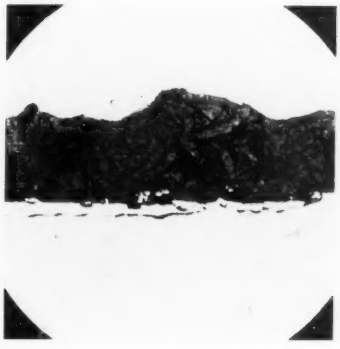


FIG. 60—Internal structure of graphite flake in ingot 21. In position of minimum reflex pleochroic absorption. Under plane-polarized light. Unetched × 1500

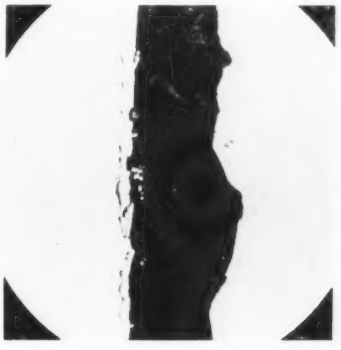


FIG. 61—Same spot as in Fig. 60, after rotating the specimen through 90°. Graphite flake in position of maximum reflex pleochroic absorption. Unetched × 1500

(Micrographs reduced to three-quarters linear in reproduction)

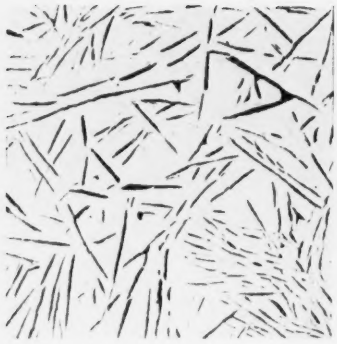


FIG. 62—Coarse graphite structure in sulphur-treated ingot (ingot 21). Unetched × 60

FIG. 63—Well-formed spherulite in ingot 20. Under plane-polarized light. Unetched × 1500

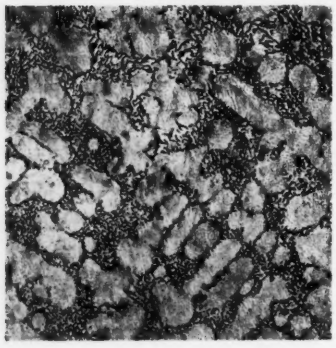


FIG. 64—General structure of material refined with ferrosilicon-zirconium addition and containing inverse chill. (*See* Fig. 65.) Undercooled graphite in pearlite. Etched with picric acid × 150

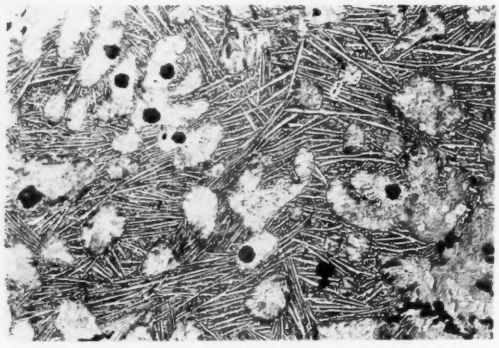


FIG. 65—An area of inverse chill in same specimen as Fig. 64. Nodular graphite in transformed austenite dendrites with laminated eutectic. Etched with picric acid × 150

FIG. 66—Detail of the white-iron structure shown in Fig. 65. Etched with picric acid × 600

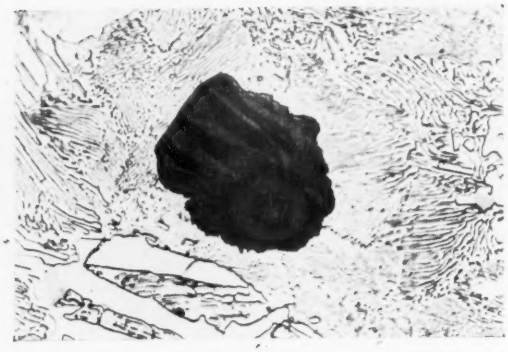


FIG. 67—Graphite nodule in region of inverse chill shown in Figs. 65 and 66. Etched with picric acid × 1500

(Micrographs reduced to three-quarters linear in reproduction)

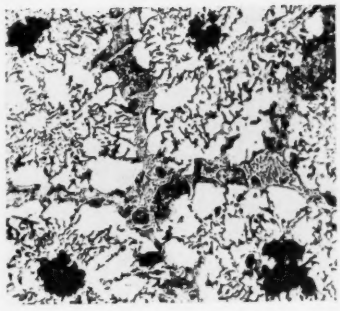


FIG. 68—Graphite nodules at centre of undercooled eutectic cells. × 200

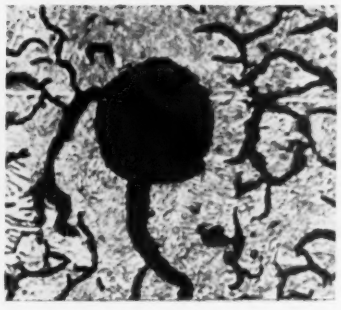


FIG. 69—Graphite spherulite with attached flake formations in hypereutectic iron. × 2000

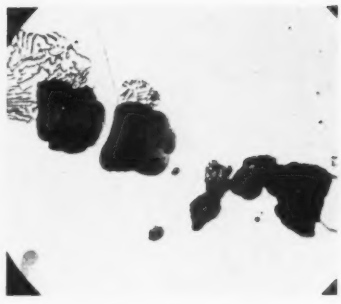


FIG. 70—Graphite spherulites in iron treated with CaSi and sulphur. Unetched × 2000

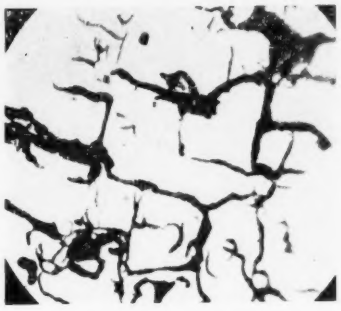


FIG. 71.—Coarse pseudo-Widmanstätten structure in iron treated with CaSi and sulphur. Unetched × 2000

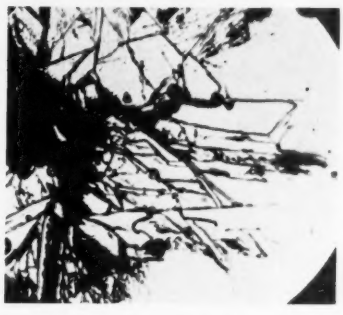


FIG. 72—Very fine pseudo-Widmanstätten structure in same sample as Fig. 71. Unetched × 2000

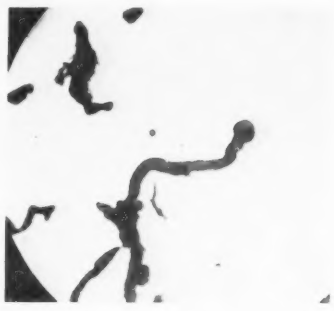


FIG. 73—Very fine spherulite formation attached to flake graphite in iron treated with CaSi and sulphur. Unetched × 2000

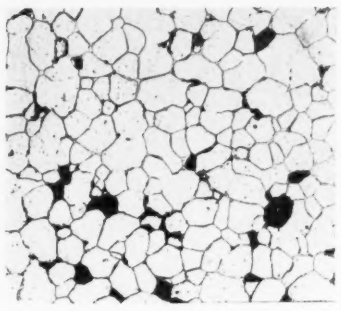


FIG. 74—Nickel–carbon alloy after attempted desulphurization with sodium carbonate. Unetched × 60

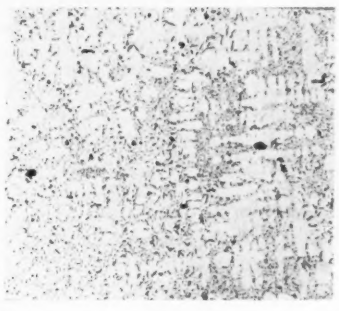


FIG. 75—Melt 31. Undercooled graphite in ingot from untreated nickel–iron–carbon alloy. Unetched × 60

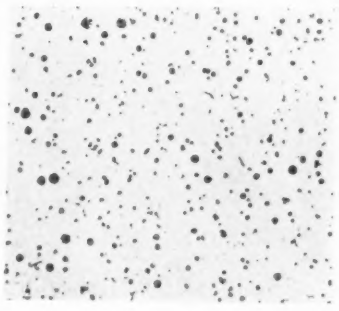


FIG. 76—Melt 31. Nodular graphite in ingot from CaSi treated nickel–iron–carbon alloy. Unetched × 60

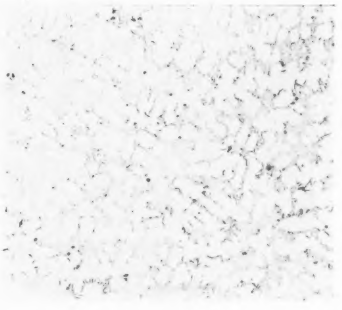


FIG. 77—Melt 32.—Undercooled graphite in ingot from untreated nickel–iron–carbon alloy. Unetched × 60

FIG. 78—Melt 32. Flake graphite in ingot from CaSi treated nickel–iron–carbon alloy. Unetched × 60

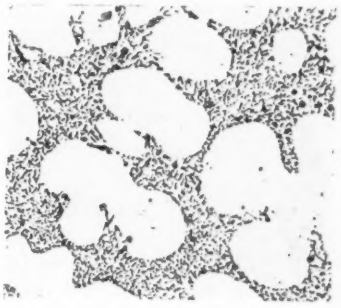


FIG. 79—Melt 33. Undercooled graphite in slowly cooled cast-iron ingot. Unetched × 150

Figs. 68 and 69 etched with picric acid

(Micrographs reduced to three-quarters linear in reproduction)

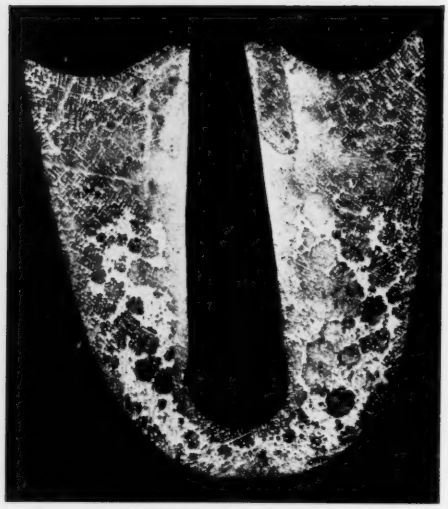


FIG. 80.—Vertical section through quenched ingot (ingot 34). Light areas are white iron. Etched with picric acid × 3½

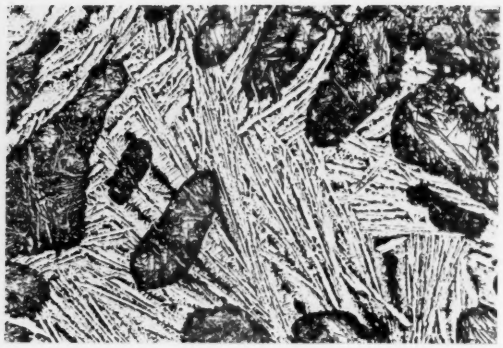


FIG. 81.—Area of white iron in quenched ingot (ingot 34). Etched with picric acid × 150

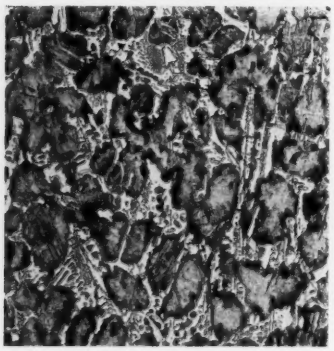


FIG. 82.—White-iron structure of ingot of same material as in Fig. 81, but with chromium addition. Etched with picric acid × 60

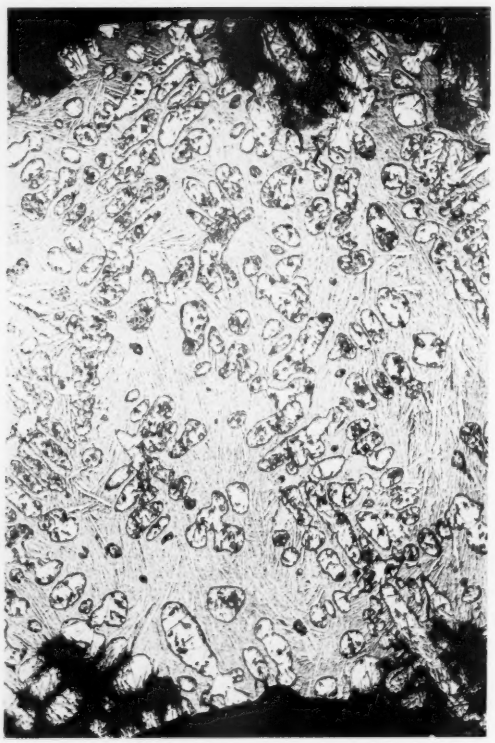


FIG. 84.—White-iron structure of ingot quenched immediately after the cessation of the eutectic arrest. Etched with picric acid × 60

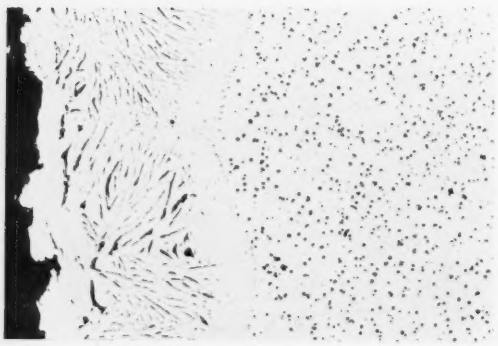


FIG. 83.—Structure of nickel-carbon ingot melted in graphite crucible. Quenched just after solidification. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

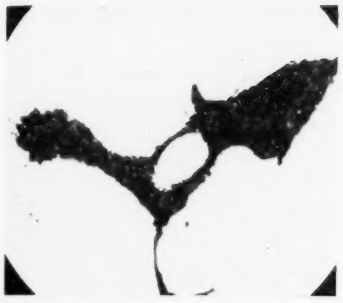


FIG. 85—Interdendritic-like graphite in ingot water-quenched from liquid condition. Unetched. × 1500

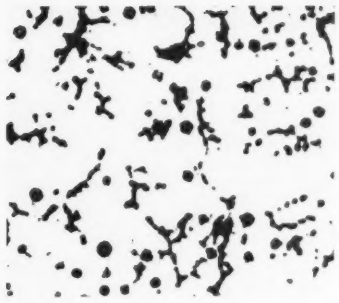


FIG. 86—Treated ingot quenched halfway through the eutectic arrest. Unetched. × 60

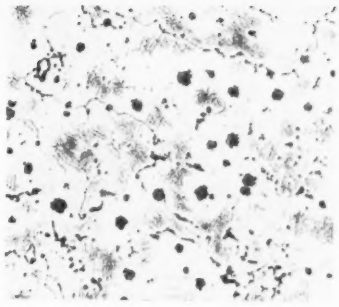


FIG. 87—Graphite nodules and fine-Widmanstätten structure of nickel carbide in water-quenched ingot. × 60

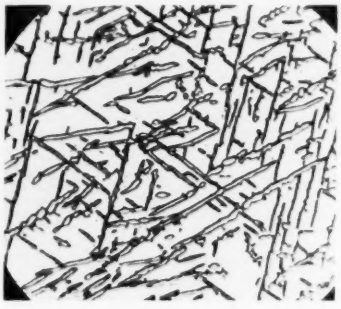


FIG. 88—Nickel-carbide Widmanstätten structure. Same sample as in Fig. 87. × 1500

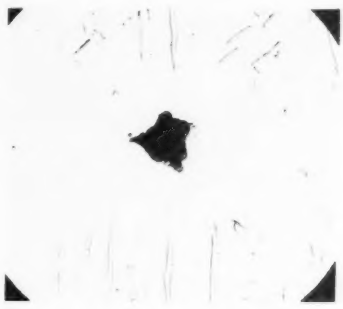


FIG. 89—Same sample as in Figs. 87 and 88, showing absence of nickel carbide in regions surrounding graphite particles. × 500

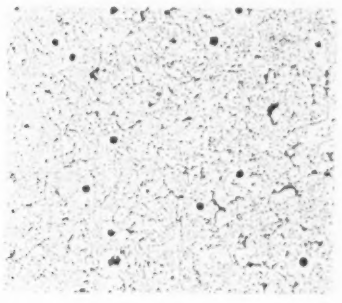


FIG. 90—Same sample as in Figs. 87 to 89, after annealing. × 60

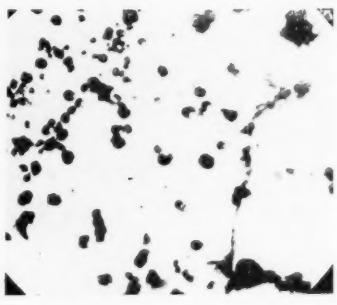


FIG. 91—Same as Fig. 90, showing fine graphite particles resulting from decomposition of nickel carbide during annealing. × 1500

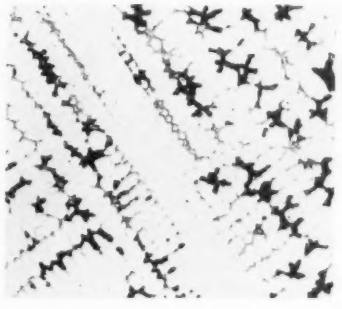


FIG. 92—Structure of quenched ingot containing approximately 60% of Fe. Carbide and intercendritic graphite. × 60

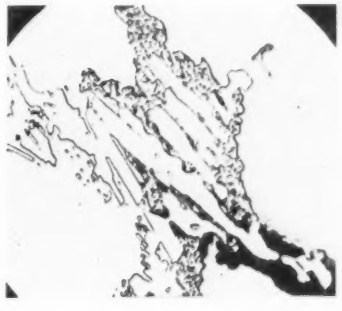


FIG. 93—Same sample as in Fig. 92, showing a eutectic carbide area. × 1500

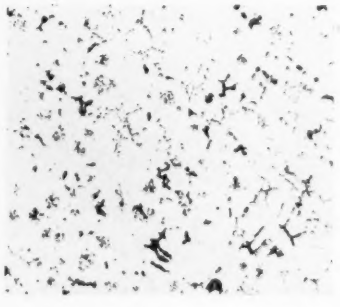


FIG. 94—Structure of quenched ingot containing approximately 40% of Fe. No carbides visible. × 60

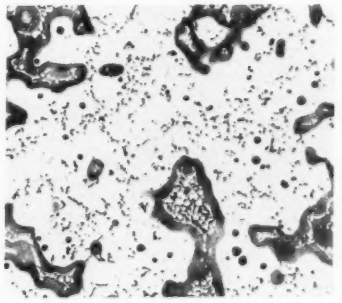


FIG. 95—Structure of CaSi treated nickel-silicon carbon ingot. × 60

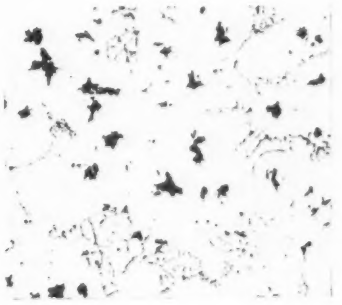


FIG. 96—Structure of Ba-Pb treated nickel-carbon ingot. Unetched. × 60

Figs. 87, 88, and 90 to 95 etched with neutral ferric chloride

(Micrographs reduced to three-quarters linear in reproduction)

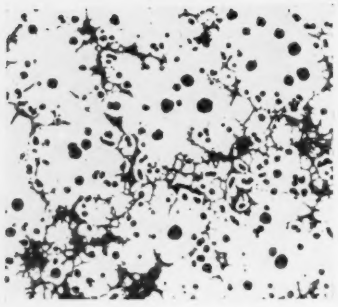


FIG. 97—Structure of Ca–Mg treated nickel–carbon ingot. × 60

FIG. 98—Same sample as in Fig. 97; $MgNi_2$ is the dispersed phase. × 1500

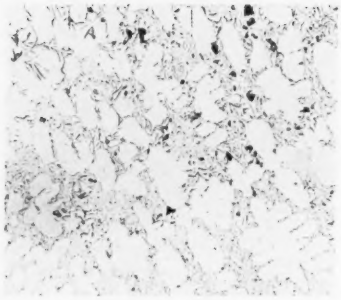


FIG. 99—Graphite structure in air-cooled cobalt–carbon melt. Unetched × 60

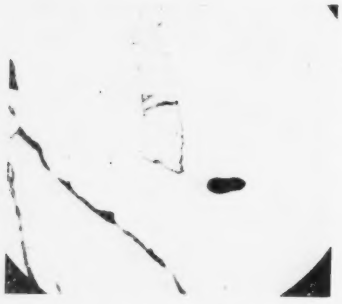


FIG. 100—Cobalt sulphide in same sample as Fig. 99. Unetched × 1500

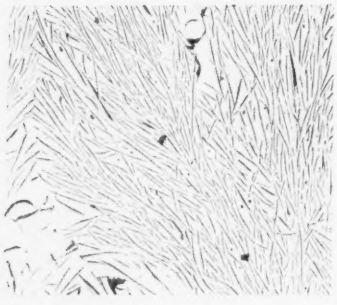


FIG. 101—General graphite structure in untreated cobalt–carbon ingot. Unetched × 60

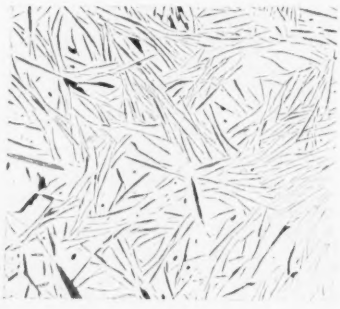


FIG. 102—General graphite structure in treated cobalt–carbon ingot. Unetched × 60

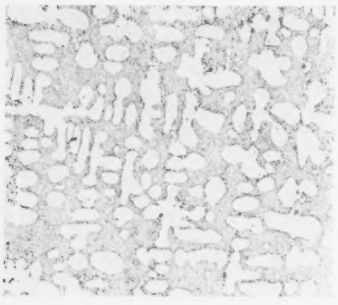


FIG. 103—Graphite area in air-cooled cobalt–carbon alloy prepared from pure cobalt powder. Unetched × 60

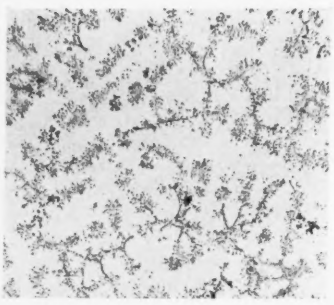


FIG. 104—Same sample as in Fig. 103, showing interdendritically arranged very fine spherulites. Unetched × 60

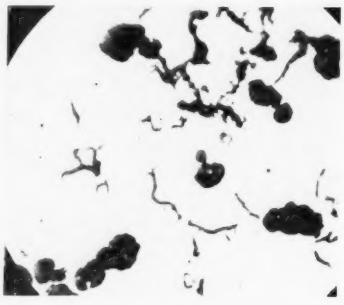


FIG. 105—Very fine spherulites in area shown in Fig. 104. Unetched × 1500

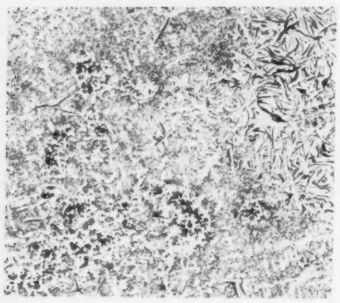


FIG. 106—Very fine undercooled graphite, some flake graphite and a few nodules in quenched cobalt–carbon ingot. × 60

FIG. 107—Structure of quenched cobalt–carbon ingot. × 60

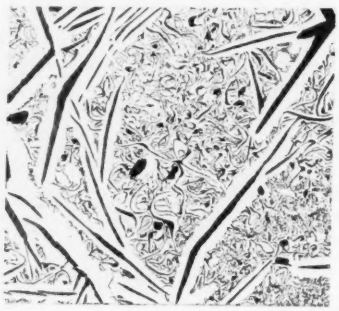


FIG. 108—Structure of quenched cobalt–carbon ingot. × 60

Figs. 97, 98, and 106 to 108 etched with neutral ferric chloride
(Micrographs reduced to three-quarters linear in reproduction)

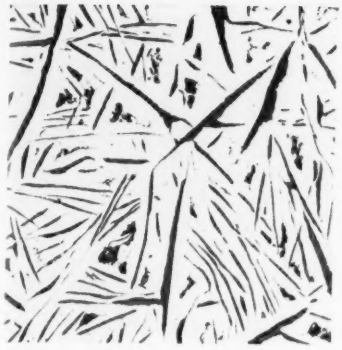


FIG. 109—Structure of quenched cobalt–carbon ingot. Etched with neutral ferric chloride × 60

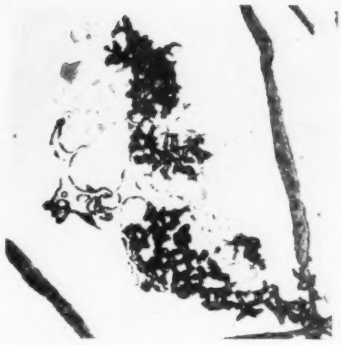


FIG. 110—Same sample as in Fig. 109, showing cobalt carbide associated with fine graphite. Etched with neutral ferric chloride × 600

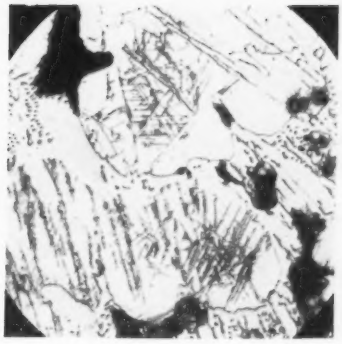


FIG. 111—Massive eutectic cobalt carbide and Widmanstätten carbide in specially treated ingot. Etched with neutral ferric chloride × 1500

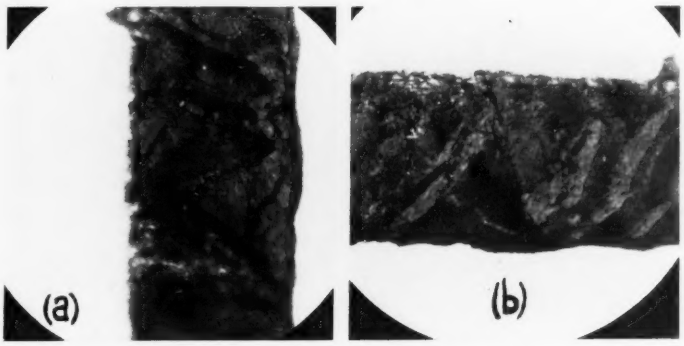


FIG. 112—Internal structure of graphite flake in a cobalt–carbon alloy: (*a*) With most of flake in position of minimum pleochroic absorption; (*b*) with most of flake in position of maximum pleochroic absorption. Under plane-polarized light. Unetched × 1500

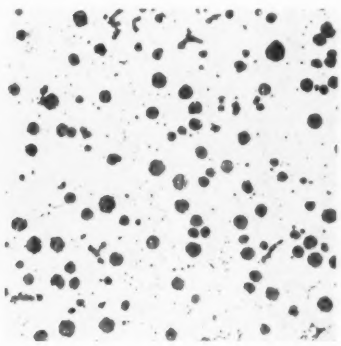


FIG. 113—Completely nodular structure of Ca–Mg treated cobalt–carbon ingot. Unetched × 60

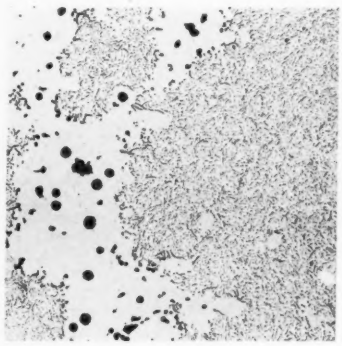


FIG. 114—Nodules and undercooled graphite in Ca–Cu treated cobalt–carbon ingot. Unetched × 60

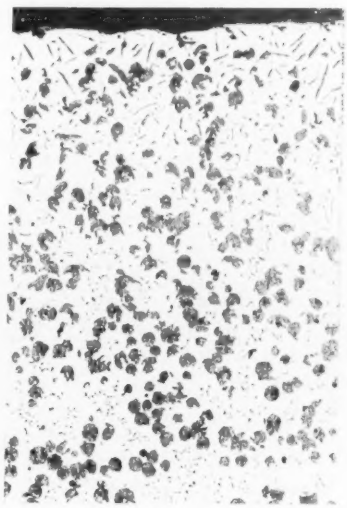


FIG. 115—Segregation of nodules at top of Ca–Cu treated cobalt–carbon ingot. Same ingot as in Fig. 114. Unetched × 60

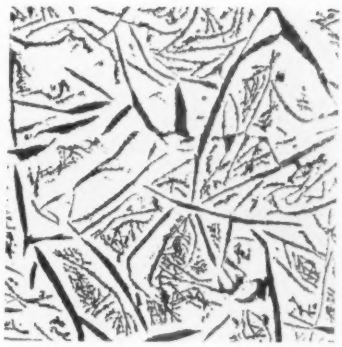


FIG. 116—Coarse graphite in Ca–Pb treated cobalt–carbon ingot. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

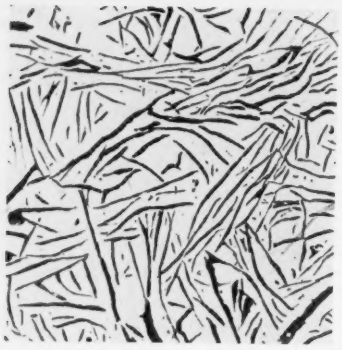


FIG. 117—Coarse flake graphite in Sr–Mg treated cobalt–carbon ingot. Unetched × 60

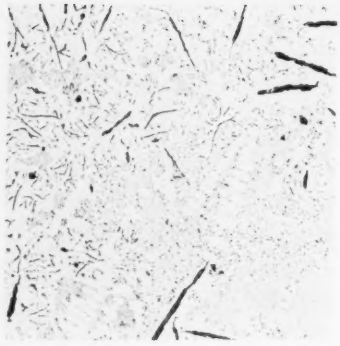


FIG. 118—Hypereutectic and fine undercooled graphite in MoS_2 treated cobalt–carbon ingot. Unetched × 60

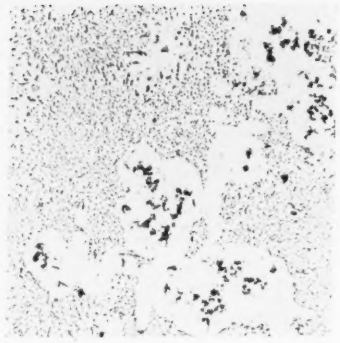


FIG. 119—"Mesh" graphite in tellurium-treated cobalt–carbon ingot. Unetched × 60

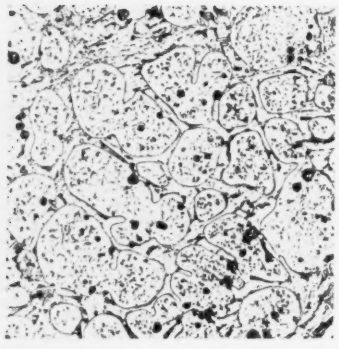


FIG. 120—Carbide and δ ferrite in quenched high-silicon cast iron. Etched with HF/HNO_3/glycerol reagent × 60

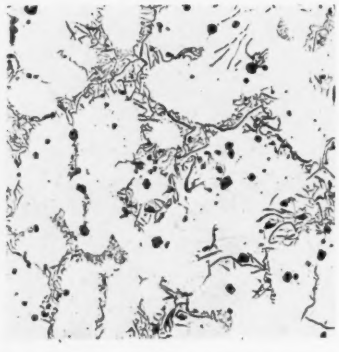


FIG. 121—Undercooled graphite + δ ferrite in quenched high-silicon cast iron. Etched with HF/HNO_3/glycerol reagent × 60

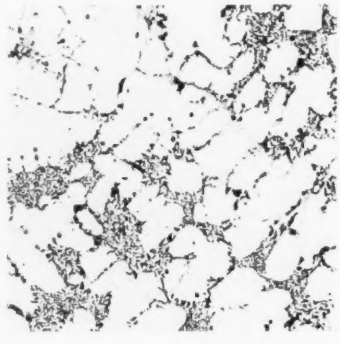


FIG. 122—Undercooled graphite and carbides in austenite matrix in good Nicrosilal casting. Etched with 10% alcoholic HCl × 60

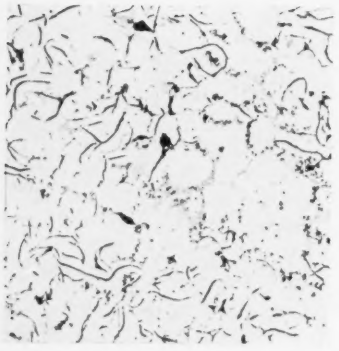


FIG. 123—Flake graphite and carbides in austenite matrix in inferior Nicrosilal casting. Etched with 10% alcoholic HCl × 60

FIG. 124—Same specimen as in Fig. 122, showing acicular dispersed carbides. Etched with 10% alcoholic HCl × 600

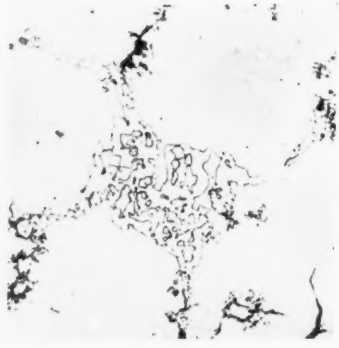


FIG. 125—Same specimen as in Fig. 123, showing relatively massive carbide. Etched with 10% alcoholic HCl × 600

(Micrographs reduced to three-quarters linear in reproduction)

calcium silicide are largely ineffectual in nucleating the melt to give flake graphite. With flake-graphite or inoculated irons solidifying at rates within the range in which flake graphite is obtained, the number of eutectic cells varies with the cooling rate. Thus within this range, and assuming that some slight undercooling takes place (most cooling curves show some slight indications of undercooling even when flake graphite is forming), then below the normal liquidus there appears to exist a range of temperature such that the number of centres or nuclei at which crystallization can begin increases with decrease in temperature. Alternatively, the nuclei become more effective at lower temperatures within this range. For cooling rates which extend the degree of undercooling beyond this temperature, a new condition applies where the number of nuclei either decrease or they become ineffectual, and in spite of an inoculating treatment the iron will solidify white or with an undercooled graphite structure, depending upon the summation of the intrinsic graphitizing and carbide stabilizing effects of the normally alloyed elements.

Influence of Hydrogen

The emphasis which Boyles placed on the effect of hydrogen in producing coarse graphite is noteworthy. Treatment with hydrogen in his experiments also increased the chilling tendency in a well-marked way, and this is a general experience. It has been argued that the effect of superheating in producing an increased chill is due to an increase in dissolved hydrogen at elevated temperatures. This involves an apparent paradox, as superheating produces undercooled graphite structures whereas hydrogen produces flake graphite structures. The hydrogen taken up by superheating would presumably originate from atmospheric moisture. Cowan [53] has demonstrated that in cupola melting the hydrogen content of the metal varies with the moisture content of the blast and that the depth of chill varies with atmospheric moisture.

Formation of Undercooled Graphite—The Work of Eash

Eash[43] has been able to show that undercooled graphite structures are formed by the decomposition of white-iron structures immediately after solidification. A bar 6 in. long by 1·2 in. in dia. was cast in a dry-sand core with a thermocouple inserted in the cavity. Cooling curves were taken on each bar cast, and, when the desired temperature had been reached, the core was broken and the specimen quenched in water. The fine graphite iron used had approximately 2·3% of total carbon and 2·6% of silicon, whilst the inoculated iron had approximately 2·1% of total carbon and 3·6% of silicon. The results obtained conclusively indicated that flake graphite irons solidify with a graphite–austenite eutectic and undercooled graphite irons solidify white, that is, with a cementite–austenite eutectic which subsequently decomposes at a temperature slightly below that indicating complete solidification. The theoretical implications of Eash's findings are immense and will be discussed in more detail later in connection with work carried out by the present authors. It is sufficient at this stage to say that, in view of the relatively crude methods used by this investigator and the somewhat abnormal compositions of his irons, the work has been repeated with the greatest care on very small cooling-curve samples for a very wide range of compositions (and for other alloys), and the findings of Eash have been fully confirmed.

Physical Properties of the Melt

Piwowarsky[41] has suggested that the variations in type of graphite structure produced by superheating may be due to variations in the molecular structure of the melt. This is a fascinating suggestion, and there are many arguments for and against the idea. Our knowledge of molecular structures in liquid metals is very limited, and the condition of carbon in cast-iron melts is unknown. Nevertheless, Schmid-Burgh, Piwowarsky, and Nipper[54] devised an ingenious apparatus to test the theory. The electrical resistivity of a ring-shaped bath of metal was measured for white and grey irons cooling from 1600° C., using a method which did not involve making any electrical connections to the melt. With grey irons a sharp increase in specific resistance was obtained during eutectic solidification which did not occur in white irons. No micrographs were given to indicate whether any of the irons had undercooled graphite, but from these results one might expect a sharp increase in resistivity immediately after solidification. The increase in resistivity of the grey irons exactly corresponded with the arrest of the accompanying cooling curve which took place at about 1130° C., which is approximately the eutectic temperature for a flake graphite iron. The investigators concluded that the results were fruitless as far as the original hypothesis was concerned. Nevertheless, the present authors feel that this method of investigation should be applied at the earliest possible moment to cast irons giving undercooled structures, and to other alloys such as aluminium–silicon, nickel–carbon, etc.

Effect of Carbon Monoxide—The Work of Schneble and Chipman

Schneble and Chipman[55] have investigated the factors involved in the superheating of grey cast iron and have attempted to explain superheating effects in terms of the gas content of the melt, particularly carbon monoxide. They constructed an induction-melting unit so that the charge could be melted and cast *in vacuo* or in any controlled atmosphere. Melts were superheated to and held at various temperatures for varying periods. In general, they found that nitrogen had little effect, that hydrogen gave flake graphite, and that carbon monoxide promoted the occurrence of undercooled graphite. This is a valuable contribution, but in attempting to explain the superheating phenomena in terms of the carbon monoxide content of the melt certain assumptions were made at variance with the known facts. They based their approach to the problem on the findings of Honda and Murakami[56] and Hayes and Scott,[57] who showed that carbon monoxide and carbon dioxide increase the graphitizing rate of solid cementite, the suggested mechanism being indicated by the following two reactions :

$$2CO = CO_2 + C \text{ (graphite)}$$
$$Fe_3C + CO_2 = 2CO + 3Fe.$$

Further, they assume that the size of the graphite flakes depends upon the stability of the carbide, so that fine graphite is produced from less stable and coarse graphite from more stable carbide. These investigators ignore the findings of Eash that the former condition results from the decomposition of solidified white iron and the other directly from the melts. The stability of cementite is known to be reduced by alloyed silicon, nickel, copper, etc., but these elements do not influence the facility with which undercooled structures can be produced by superheating. The fact that superheating also increases the tendency to chill was not referred to. That carbon monoxide can be an important factor in superheating is undeniable, but the theory of these investigators seems untenable. They give no indication of why titanium and zirconium promote undercooling without any particular superheating.

The very frequent occurrence of ferrite with undercooled graphite seems to present itself as a problem according to some references, and Piwowarsky[58] has erected some very improbable theorizing to explain its presence. He assumed this ferrite to be formed immediately after solidification. The present investigators have quenched large numbers of samples with fine undercooled graphite immediately after solidification and from lower temperatures, and have never found any ferrite to occur until the critical range has been reached. It is not generally appreciated that the state of division of graphite arising at the eutectic has a very pronounced effect in determining whether the eutectoid transformation will take place according to the stable or the metastable system, and upon the subsequent stability of any pearlite which forms on exposure to sub-critical temperatures. Taking an extreme case, an iron containing 4% of silicon and 2·8% of total carbon can be made completely ferritic when the graphite is entirely of the fine undercooled variety, and completely pearlitic when the graphite is in the normal flake form.

Carbide Stabilization

Considerable obscurity centres around the reasons why some elements are carbide stabilizers and others graphitizers. These problems are relevant to the present discussion, and a statement of the more obvious points is made below. No satisfactory explanation is known for the observed facts.

Silicon—A graphitizing element. Up to about 6% silicon progressively reduces the stability of cementite. In the range 12–14% of silicon the carbide which occurs seems to possess a stability higher than one would expect if the effect was merely an extrapolation of that at lower silicon contents. A very stable carbide (SiC) is known, soluble in cast-iron melts.

Nickel—A mildly graphitizing element. Will dissolve in cementite to give $(Fe,Ni)_3C$.[59] Ni_3C is probably very unstable. (This will be dealt with later.)

Copper—Mildly graphitizing. Similar in effect to nickel. No information about solubility in Fe_3C.

Aluminium—Graphitizing when added as a ladle addition in small amounts (less than 0·1%). This is probably owing to a deoxidizing effect. Graphitizing when added as an alloying element in amounts between 0·5 and 3%. Carbide-stabilizing in amounts ranging from 5 to 9%, probably owing to the formation of an iron–aluminium carbide.

Manganese—A carbide stabilizer. $(Fe,Mn)_3C$ probably formed. $Mn_{23}C_6$ and Mn_7C_3 have been reported in alloys with more than 80% of manganese.[60]

Chromium—A carbide stabilizer. $(Fe,Cr)_3C$ probably formed. Gives defined stable carbides. Cr_4C and Cr_3C_2, and $(Cr,Fe)_4C$, $(Cr,Fe)_7C_3$, and $(Cr,Fe)_3C_2$ probably formed.[61]

Molybdenum — A carbide stabilizer. $(Fe,Mo)_3C$ probably formed. Gives a stable carbide, probably Mo_2C, and $(Fe,Mo)_{23}C_6$ has been reported in steels.[62]

Vanadium—A carbide stabilizer. $(Fe,V)_3C$ probably formed. Gives stable vanadium carbides.

Tungsten—Not clearly defined. May be mild carbide stabilizer. Forms stable tungsten carbides WC and W_2C.

Boron—A carbide stabilizer. Forms a stable boron carbide.

Sulphur—As iron sulphide, sulphur is a carbide stabilizer. Combination with cementite is not proven. As manganese sulphide it behaves neutrally.

Tellurium—Exists as FeTe and $MnTe_2$. Appears to be a powerful carbide stabilizer in either form.

Titanium—A graphitizer when added in small amounts (less than 0·1%); this may be a deoxidizing effect resembling that of aluminium. Forms a very stable carbide (TiC), but behaves neutrally to cementite as an alloying element.

Zirconium—Behaves very similarly to titanium. Forms a stable carbide (ZrC).

Bismuth—A carbide stabilizer, in which respect it behaves similarly to tellurium, giving the same sort of rimmed chill.

It is understood that iron carbide may not be thermodynamically stable, but conventional nomenclature uses "carbide stabilization" to indicate an influence increasing the persistence of cementite, and "graphitizer" to indicate an influence decreasing its persistence. Those elements (Mn, Cr, Mo, V, and probably W) which form stable carbides and whose atoms may behave vicariously with Fe in Fe_3C, are carbide stabilizers. The action of the graphitizers (silicon, copper, and nickel) is unknown. The carbide stabilizers sulphur, selenium, and tellurium present a separate group, and we might expect a common explanation for all three. The effect of selenium has not yet been very thoroughly explored, but there are significant differences between the effects of sulphur and tellurium. Sulphur behaves as a carbide stabilizer only when in the form of FeS, but tellurium exerts this effect when present as either FeTe or $MnTe_2$ (the reasons for assigning these formulæ will be dealt with in a separate paper). Sulphur in the form of MnS behaves neutrally to cementite. The powerful effect of tellurium no doubt prompted its commercial application.[63] In addition to this difference, tellurium imparts a characteristic appearance to the chill, *viz.*, a very sharp line of demarcation between white and grey zones. Morgan, working in the British Cast Iron Research Association laboratories, found the chilling effect of tellurium to be very sensitive to pouring temperature.[64] Fig. 27 shows the effect of pouring temperature on the fracture of an iron containing tellurium. It might be suggested that this apparent influence of tellurium was due to its gradual loss from the melt. This is disproved by the fractures illustrated in Fig. 28. In this case an iron treated with tellurium was remelted in a crucible furnace and then poured at varying temperatures. The same effect will again be noted. There undoubtedly has been some loss of tellurium during remelting, but even after this lengthy process sufficient has remained to give a marked effect and to cause the sharp chilled rim to return with the low pouring temperature. The effect of pouring temperature on a characteristic sulphur chill is shown in Fig. 26. Figs. 26, 27, and 28 were all taken from hitherto unpublished work by Morgan, and all show the fractures of 1·2-in. test-bars poured in green sand. It is quite obvious from these illustrations that, whatever factors cause chill depth to increase with increasing pouring temperature, tellurium increases the sensitivity of the iron to these influences. In sulphur and tellurium a very complex case of carbide stabilization exists. From work not yet complete, it appears that bismuth may behave in a manner similar to that of tellurium.

The graphitizing influence of aluminium, titanium, and zirconium, when added in small amounts immediately before pouring, may be related to their "deoxidizing" effects. There is no sure foundation for this assumption, but on the face of it it seems to be the only property which they all possess in common. There is ample ground here for investigation of the possibility of oxygen content and the method of its fixation influencing chilling characteristics. The advantages occurring from the use of aluminium along with a ferrosilicon inoculant and of the use of a ferrosilicon containing manganese and zirconium (S.M.Z.) have been referred to previously. A significant fact in this context is that whilst these two give, on the average, higher-strength grey irons than those obtained by using ferrosilicon alone, no differences in the graphite or matrix structure of grey material can be observed (irons treated with S.M.Z. always contain the characteristic zirconium sulphides). The authors have no leading observations to make upon the carbide-stabilizing effect of boron. Boron added in small amounts has been reported[65] to have a graphitizing effect when added to

malleable iron. This effect has also been noted in the laboratories of the British Cast Iron Research Association when ferroboron was used for the alloy addition. Ferroboron usually contains large amounts of aluminium or other deoxidizers, and it is suggested that these are responsible for the graphitizing effects. No such graphitizing effect has been noted when boron carbide was used to make the addition to grey and mottled irons. The effect of alloyed aluminium is anomalous, but it should be noted that an aluminium carbide (Al_4C_3) can be formed.

The industry will recognize other influences on carbide stabilization than those listed above. As a general rule, all other conditions being the same, the tendency to chill increases with decreasing total carbon content. Other effects, all very familiar to practical foundrymen, may be related to the oxygen content of the melt. This will include the effect of slag composition and melting stock. Compositions being equal, the chilling tendency and graphite structures of irons prepared from hematite and steel on the one hand, and refined iron on the other, will differ. Similarly, electric-furnace, crucible-melted, and cupola-melted irons may vary in structure and chilling tendency. On the whole, cupola irons will show a coarser graphite structure with less tendency to give undercooled structures than is the case with electric-furnace and crucible-melted irons. It can be argued that this is owing to an inoculating effect obtained by the metal being in contact with coke. On the other hand, the atmospheres in these melting furnaces are quite different. The practical foundryman also reports that the chilling tendency of an iron prepared by mixing molten high-silicon and molten low-silicon irons differs from that of a single melt of the same final silicon analysis. Many instances of this type obtrude upon the problem, but few have been stated with sufficient precision to enable them to be considered in detail in the present discussion.

Effect of Slag Composition

Particularly revealing were the experiments of Bardenheuer and Reinhardt,[95] who found that melting under mill scale in a high-frequency induction furnace tended to produce flake graphite, and melting under glass tended to produce undercooled graphite. The whole field of slag–metal reactions for cast iron is practically unexplored, but this example suffices to show the importance of its consideration. The acidity or basicity of the slag may determine the FeO or SiO_2 content of the melt, which may in turn affect the tendency to undercool. On the other hand, it could be argued that high-silica slags react with the carbon of the melt to give carbon monoxide and silicon. The silicon would be taken up by the melt and, according to Schneble and Chipman, the carbon monoxide would facilitate undercooling. These suggestions, like many others in this field, are purely speculative.

Inverse Chill

Bound up with the mechanism of formation of undercooled graphite structures is the fact that inverse chill occurs most frequently in undercooled-graphite-containing irons. (Inverse chill refers to the fracture or microstructure which arises when the central and more slowly cooled zones of a section have a white-iron structure and the outer and more rapidly cooled zones have a grey-iron structure. This is the reverse of what one would normally expect.) No satisfactory explanation has been given for inverse chill, but this connection must throw some light on its formation. The present authors distinguish two varieties of inverse chill. In irons which have completely fine undercooled graphite structure, inverse chill occurs in a wide variety of degree, but in every case with a characteristic appearance. In small amounts it has the appearance of an undissolved alloy addition, and under the microscope appears as indicated in Fig. 29. Note should be made of the acicular character of the carbides in this micrograph. Fig. 30 indicates typical fractures of bars with undercooled graphite and central areas of inverse chill. In a later section of this paper this matter is dealt with again. Another type of inverse chill occurs when the grey portions of the melt are not necessarily of the undercooled structure. In these cases the outer grey portions are generally only a very thin band (Fig. 31). It is suggested that this is an inoculating effect of the mould face. In a similar manner, when deliberately producing irons with undercooled graphite structures, a fine rim of normal flake graphite has been found at the edges of the test-bars, again apparently owing to the silica sand or coal dust at the mould face inoculating the casting during solidification. In another context, silica will be shown to have an inoculating effect, and, using the theory of Schneble and Chipman, it is conceivable that under these conditions silicon reacts with the carbon of the melt to give carbon monoxide, which in turn facilitates graphite formation. In this latter case the effect would not be a true inoculating effect.

Reference has been made to the carbide-stabilizing effect of bismuth, but this element exercises another influence of interest; it has

a pronounced graphite-refining effect. Treatment of a melt with bismuth is as effective as titanium in replacing flake graphite structures with undercooled graphite structures, even without any subsequent treatment of the melt with carbon dioxide. Figs. 32 and 33 illustrate this effect in 1·2-in. test-bars, the untreated metal having the following analysis :

T.C.	Si	Mn	S	P
3·18%	1·89%	0·63%	0·034%	0·045%

An addition of 0·2% of bismuth was made to the iron shown in Fig. 33, but chemical analysis gave only 0·02%.

Summary of Graphite-Formation Anomalies

A detailed account has now been given of the more important anomalies of graphite formation in grey cast iron, and it is appropriate at this stage to recapitulate with a short summary. The outstanding factors needing explanation, correlation, or proof may be listed as follows :

(1) Formation of undercooled structures.
(2) Production of undercooled structures by superheating.
(3) Production of undercooled structures by titanium and zirconium, with or without carbon dioxide.
(4) Effect of carbon dioxide in (3).
(5) Production of undercooled structures by bismuth.
(6) Titanium and zirconium refine only when present as TiC and ZrC.
(7) Von Keil's silicate-slime theory.
(8) Norbury and Morgan's theory.
(9) Removal of porosity by undercooled graphite structures.
(10) Inoculating effect of ferrosilicon and calcium silicide.
(11) Inoculating effect of graphite.
(12) Improvement in ferrosilicon inoculation by inclusion of powerful deoxidizers such as zirconium or aluminium.
(13) Explanation of undercooling according to classical theory.
(14) Simultaneous occurrence of primary dendrites from both sides of eutectic.
(15) Inoculating effect of kish.
(16) Theory of superheating effect being due to elimination of graphite nuclei (Massari and Lindsay).
(17) Rapid solution of graphite above the liquidus (Piwowarsky).
(18) Inoculation effect of ferrosilicon owing to the concentrations of silicon giving graphite (Eash).
(19) Formation of undercooled structures by decomposition of a white iron (Eash).
(20) Effect of hydrogen : Flake graphite and carbide stabilizer.
(21) Vacuum melting : Fine undercooled graphite.
(22) Effect of sulphur : Flake graphite.
(23) Large eutectic cell size of undercooled structures.
(24) Flake graphite forming directly from melt.
(25) Effect of moisture in cupola blast in increasing chill.
(26) Effect of carbon monoxide (Schneble and Chipman).
(27) Effect of increasing pouring temperature in increasing chill.
(28) Effect of superheating in increasing chill.
(29) Carbide-stabilizing action of chromium, manganese, carbon, molybdenum, and tungsten.
(30) Carbide-stabilizing action of sulphur, selenium, and tellurium.
(31) Graphitizing action of silicon, nickel, and copper.
(32) Graphitizing action of ladle additions of aluminium, titanium, and zirconium.
(33) Carbide-stabilizing action of bismuth.
(34) Carbide-stabilizing action of boron.
(35) Aluminium first a graphitizer (up to about 3%), then a carbide stabilizer.
(36) Sharp chill with tellurium, and its sensitivity to pouring temperatures.
(37) Formation of inverse chill.
(38) Effect of melting unit.
(39) Effect of nature of melting stock.
(40) Refining action on graphite of copper in excess of liquid miscibility.
(41) Effect of slag composition.

Experiments on the Nickel–Carbon Alloys

Object of this Work

At the outset these experiments were motivated by the feeling that the findings of Eash that undercooled fine graphite structures were formed by the decomposition, immediately after solidification, of a white-iron structure were the most important in recent years. With this knowledge it was obvious that the picture of undercooling based on our knowledge of the betol–salol system was at the best limited, and certainly unsatisfactory. In the investigations on undercooling in this system there is no reference to the formation of metastable compounds formed between betol and salol at the hypertectic temperature. Therefore the question arose as to whether the mode of undercooling in iron–carbon–silicon alloys and cast irons was unique, or whether

this type of solidification could be reproduced for other systems, by the study of which we might be led closer to the truth about cast irons. This course was further favoured in view of the manner in which the form of the binary equilibrium diagrams of two elements with a third element can be related to their positions in the periodic classification of elements. There is frequently considerable analogy between two systems, *AB* and *AC*, when *A* is an element common to both, and *B* and *C* are elements of the same group in the periodic table of elements. Useful inferences about the system *AB* may be obtained from a study of the system *AC*, and *vice versa*. Inferences of this type must, of course, be accepted with extreme caution, and the authors have tried to preserve this cautious attitude throughout the following experiments.

When considering the problem of graphite formation in iron alloys, the value of studying the nickel–carbon and cobalt–carbon systems came immediately to mind. Iron, cobalt, and nickel are neighbours in the first long period and in this period they are the last three transitional elements. Their atomic numbers are 26, 27, and 28, and their atomic weights are 55·85, 58·94, and 58·69, respectively. The increase in electrons consequent upon increasing atomic number takes place by the addition of electrons to the inner orbits.

With carbon, these three elements all yield equilibrium diagrams of great similarity. The systems are eutectiferous, involving a eutectic of graphite and a solid solution of carbon in the metal. The crystal structure of these solid solutions is probably the same in all three cases The liquidus on the metal-rich side of this eutectic corresponds to the deposition of this solid solution from the melt, and on the carbon-rich side of the eutectic it indicates the solubility of graphite in the melt. In every case the solubility of carbon in the solid solution decreases with decreasing temperature.

It was decided to investigate the nickel–carbon alloys first. This decision was taken largely because a supply of reasonably pure nickel was readily available. In addition, this element is a much more important alloying element in cast iron than is cobalt. The nickel–carbon system has been investigated on several occasions,[67–70] and there is fair agreement in the results obtained. The diagram is given in Fig. 126, in which all the available data have been included. Ruff and Martin[68] claimed to have obtained the carbide Ni_3C, indicated by the break on the liquidus curve, by quenching from a high temperature. Their micrographs give no positive evidence of

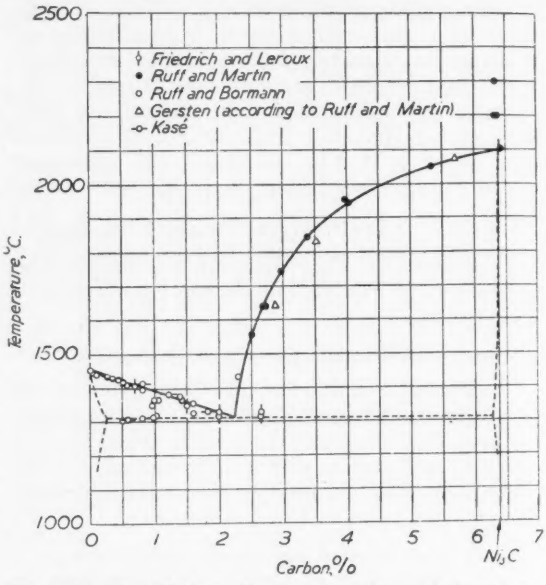


FIG. 126—Equilibrium diagram for the nickel–carbon alloys (Marsh)

this compound. The general inference to be made from this early work is that if a nickel carbide exists, it is extremely unstable and that all alloys with carbon in excess of solid solubility consist of solid solution + graphite. Jacobson and Westgren[71] prepared Ni_3C by carburizing nickel powder in carbon monoxide at approximately 285° C. for 24 to 168 hr. Their X-ray analysis of this compound showed it to be hexagonal close-packed and hence quite different from cementite.[97,98] The nickel carbide Ni_3C would appear therefore to be very unstable at high temperatures, but at least to be persistent at low temperatures.

First Series of Experiments on Nickel–Carbon Alloys

The first series of experiments on nickel–carbon alloys were all carried out in a small high-frequency induction furnace, using a fireclay crucible holding about 8 lb. of metal. The charges in every case consisted of pure nickel shot and electrode carbon. For micro-examination and chemical analysis small ingots 1·2 in. in dia. and 2–6 in. long were cast in green-sand moulds. The initial object of these experiments was to determine whether nickel–carbon alloys reacted to various treatments in a manner similar to that of cast irons. A large number of melts were carried out in this series, but, for the

sake of brevity, only the more important melts indicating some definite fact will be described. Similarly, every sample cast was examined microscopically and photographed, but here only the important relevant features will be illustrated by micrographs.

Melt No. 1—This was of hypo-eutectic composition, and the charge contained 2% of carbon. The metal bath was superheated to 1600° C. and the first ingot, 1*A*, was poured. Calcium silicide (0·5%) was then added to the melt and ingot 1*B* was poured. The melt was then treated with ferrosilicon-zirconium* (0·5%) and ingot 1*C* was poured.

Ingot 1*A* had a graphite structure exactly similar to the undercooled structure of a hypo-eutectic cast iron. The actual structure is shown in Fig. 34.

Ingot 1*B* had a structure consisting entirely of spherulitic carbon nodules. This structure is illustrated in Fig. 35. This sample resembled the structure of an annealed malleable iron, and its occurrence in a cast ingot without any heat-treatment was considered a remarkable occurrence. Ingot 1*C* had a structure consisting entirely of the fine undercooled type of graphite as found in ingot 1*A*. This structure is illustrated in Fig. 36.

Quite obviously the results of this melt indicated that superheating had a similar effect in nickel–carbon alloys to that in cast iron. Calcium silicide additions had a very different effect, and ferrosilicon-zirconium probably behaved as in cast irons.

Melt No. 2—This was carried out originally to investigate the effect of adding a cementite stabilizer to a nickel–carbon alloy. It was fully appreciated that because an element may stabilize iron carbide in iron–carbon–silicon alloys, it does not therefore necessarily have the same effect on carbides in other systems, in this case Ni_3C in nickel–carbon alloys. Examples of this kind are common in cast irons themselves; for instance, silicon has no apparent effect on the stability of titanium carbide, and chromium-containing carbides in the austenitic iron Nicrosilal are remarkably persistent in spite of the simultaneous presence of 17–20% of nickel and 4–5% of silicon, the combined effects of which should easily neutralize the carbide-stabilizing influence of 2–4% of chromium, according to the popularly accepted relative values of these influences. Many other examples can be quoted, and these instances might assist in a better understanding of the reasons for the observed effects of these elements. In this instance tellurium was chosen.

The charge was exactly the same as that used for melt No. 1. After superheating, ingot 2*A* was poured. This gave an undercooled graphite structure exactly similar to 1*A*. Tellurium (0·2%) was then added and ingot 2*B* was poured; the melt was then treated with 0·6% of ferrosilicon-zirconium and ingot 2*C* was poured. Ingots 2*B* and 2*C* both contained coarse flake graphite structures. The structure of the tellurium-treated ingot is shown in Fig. 37. This coarse graphite structure was exactly similar in appearance to its counterpart in grey cast iron. Under these conditions and in the presence of tellurium, the ferrosilicon-zirconium did not appear to have any effect. After treatment with tellurium, small amounts of a minor phase, probably nickel telluride (NiTe) [72] were observed in the structure. No carbides were seen.

Melt No. 3—This melt was carried out to determine whether the nodular structure produced by calcium silicide would persist after treatment with tellurium. The charge used was the same as those used previously. After superheating to 1600° C., an ingot, 3*A*, was poured; an addition of calcium silicide was then made and ingot 3*B* was poured. Finally, tellurium was added and ingot 3*C* was poured.

Ingot 3*A* was exactly similar to ingot 1*A*, showing the undercooled graphite type of structure (*see* Fig. 34). Ingot 3*B* had the graphite in the spherulitic form as in Fig. 35, and ingot 3*C* had coarse flake graphite as in Fig. 37.

These results indicate the predominating influence of the tellurium over the calcium silicide addition.

Melt No. 4—As tellurium has, by virtue of its position in the periodic table, similar properties to those of sulphur, and as the latter element is more important as a constituent in cast iron, it was decided to investigate its effect when present as nickel sulphide and when present as manganese sulphide.

The charge employed was similar to those used previously. After superheating to 1600° C., ingot 4*A* was poured; 0·3% of roll sulphur was then added and ingot 4*B* was poured; 1% of manganese as ferromanganese (80%) was then added and ingot 4*C* was poured.

Ingot 4*A* had the undercooled graphite structure (*see* Fig. 34). Ingot 4*B* had a coarse flake graphite structure, and ingot 4*C* was similar to 4*B*, but with slightly finer graphite. These two structures are illustrated in Figs. 38 and 39, respectively.

* It has already been shown in an earlier section that this addition causes the refinement of the graphite structure in cast iron.

The nickel sulphide, Ni_3S_2[47,73-75] was visible in ingot 4*B*, and only manganese sulphide was seen in ingot 4*C*.

In this case an obvious analogy exists between nickel–carbon alloys and iron–carbon–silicon alloys. It will be recalled from an earlier section that Boyles had demonstrated that sulphur additions to iron–silicon–carbon alloys causes the formation of flake graphite structures.

Melt No. 5—This melt was carried out to investigate the effect of hydrogen on superheated metal. A similar charge to that used previously was superheated to 1600° C. and ingot 5*A* was poured. The melt was held at this temperature, hydrogen was bubbled through for 2 min., and ingot 5*B* was poured.

The structure of ingot 5*A* is shown in Fig. 40 and consisted of a mixture of the fine undercooled variety and a coarser form. Ingot 5*B* had a uniform graphite structure rather coarser than the typical undercooled variety reported previously. This structure is illustrated in Fig. 41.

Melt No. 6—To investigate the effect of hydrogen treatment upon a melt already treated with calcium silicide, a charge similar to those previously used was superheated to 1600° C. and ingot 6*A* was poured ; calcium silicide was then added and ingot 6*B* was poured. The melt was then kept at temperature, hydrogen was bubbled through for 2 min., and ingot 6*C* was poured.

Ingot 6*A* showed the familiar undercooled graphite structure. Ingot 6*B* had all the graphite in the nodular form (as in Fig. 35), with the exception of a rim of undercooled graphite (Fig. 43). Ingot 6*C* had a structure consisting entirely of fine undercooled graphite, as illustrated in Fig. 42. Obviously, the nodular structure produced by calcium silicide can be removed by hydrogen. The undercooled graphite rim on ingot 6*B* will be referred to later.

Melt No. 7—Again the charge was similar to the previous melts. After superheating to 1600° C. ingot 7*A* was poured. Tellurium was then added and ingot 7*B* was poured ; hydrogen was then bubbled through the melt for 2 min. and ingot 7*C* was poured. Ingot 7*A* had an undercooled graphite structure and ingots 7*B* and 7*C* had coarse flake graphite structures. Hydrogen therefore had no effect on the coarsening influence of tellurium.

Melt No. 8—The charge in this case contained $3\frac{1}{4}\%$ of carbon, that is, sufficient to render the melt hypereutectic. After superheating to 1600° C. ingot 8*A* was poured ; calcium silicide was then added and ingot 8*B* was poured.

Ingot 8*A* had long straight " kish " or hypereutectic graphite flakes with an undercooled graphite eutectic, together with primary dendrites from the metal-rich side of the eutectic (Fig. 45). This may be taken to indicate undercooling of the type described previously and indicated in Fig. 7*c*. Ingot 8*B* had a structure consisting entirely of spherulitic carbon nodules (Fig. 44). It would appear that both hypereutectic and eutectic graphite can be caused to form in spherulites by the calcium silicide treatment of the melt.

Melt No. 9—The charge for this melt contained $3\frac{1}{4}\%$ of carbon, as for melt No. 8. After superheating, ingot 9*A* was poured, tellurium was then added, and ingot 9*B* was poured.

Ingot 9*A* showed areas very similar in structure to ingot 8*A*, but considerable flotation of kish graphite to the top regions of the ingot had taken place. As a result, towards the upper central zones of the melt the structure was wholly eutectic (Fig. 46), whilst at the top the graphite was much coarser and some well-formed spherulites were to be seen (Fig. 47). It would appear that hypereutectic graphite can cause inoculation of the solidifying ingot and also that this graphite can give spherulites without additions of calcium silicide. Ingot 9*B*, treated with tellurium, had a uniformly very coarse graphite structure (Fig. 48).

Melt No. 10—A charge containing 2% of carbon, that is, hypo-eutectic, was superheated to 1600° C. and ingot 10*A* was poured. Titanium (0·5%) as ferrosilicon-titanium was then added, and, when dissolved, carbon dioxide was bubbled through the melt for 2 min. ; ingot 10*B* was then poured.

Ingot 10*A* had a fine undercooled graphite structure (Fig. 49) and ingot 10*B* had a similar, but coarser graphite structure (Fig. 50).

Melt No. 11—A charge containing 2% of carbon was superheated to 1600° C. and calcium silicide was added. Ingot 11*A* was then poured. Titanium (0·5%) was then added, carbon dioxide was bubbled through the melt for 2 min., and then ingot 11*B* was poured.

Ingot 11*A* had its graphite in nodular form as illustrated previously, but in ingot 11*B* a graphite structure similar to that illustrated in Fig. 50 had formed.

Melt No. 12—A charge containing 2% of carbon was superheated to 1600° C. and then 0·5% of sulphur, as roll sulphur, was added and ingot 12*A* was poured. The melt was then treated with calcium silicide and ingot 12*B* was poured.

Ingot 12*A* and ingot 12*B* both had fairly coarse flake graphite structures. In the presence of

TABLE I—*Summary of First Series of Nickel–Carbon Melts*

Ingot No.	Charge Composition	Treatment	Structure
		Melt No. 1	
1*A*	2% C	Superheated	Undercooled
1*B*		CaSi	Nodular
1*C*		FeSiZr	Undercooled
		Melt No. 2	
2*A*	2% C	Superheated	Undercooled
2*B*		Te	Coarse
2*C*		FeSiZr	Coarse
		Melt No. 3	
3*A*	2% C	Superheated	Undercooled
3*B*		CaSi	Nodular
3*C*		Te	Coarse
		Melt No. 4	
4*A*	2% C	Superheated	Undercooled
4*B*		S	Coarse
4*C*		Mn	Coarse
		Melt No. 5	
5*A*	2% C	Superheated	Undercooled
5*B*		H_2	Coarser
		Melt No. 6	
6*A*	2% C	Superheated	Undercooled
6*B*		CaSi	Nodular
6*C*		H_2	Coarser
		Melt No. 7	
7*A*	2% C	Superheated	Undercooled
7*B*		Te	Coarse
7*C*		H_2	Coarse
		Melt No. 8	
8*A*	3¼% C	Superheated	Kish and undercooled
8*B*		CaSi	Nodular
		Melt No. 9	
9*A*	3¼% C	Superheated	Kish and undercooled
9*B*		Te	Coarse
		Melt No. 10	
10*A*	2% C	Superheated	Undercooled
10*B*		Ti + CO_2	Coarser
		Melt No. 11	
11*A*	2% C	Superheated + CaSi	Nodular
11*B*		Ti + CO_2	Coarse undercooled
		Melt No. 12	
12*A*	2% C	S	Coarse
12*B*		CaSi	Coarse
		Melt No. 13	
13*A*	2% C	S + Mn	Coarse
13*B*		CaSi	Coarse
		Melt No. 14	
14	2% C	Si + Mn + P + S	Coarse

sulphide, calcium silicide has no tendency to cause the formation of nodular structures.

Melt No. 13—A charge was melted as for melt No. 12, 0·5% of sulphur as roll sulphur was added, then 1·0% of manganese as 80% ferromanganese, and ingot 13*A* was poured. Calcium silicide was then added and ingot 13*B* was poured.

Both of these ingots had coarse flake graphite structures. Calcium silicide therefore does not give nodular structures in the presence of manganese sulphide.

Melt No. 14—The charge for this melt was adjusted to give a composition resembling that of a cast iron, but based on nickel instead of iron. It consisted of pure nickel shot, 2% of electrode carbon, 2·5% of silicon as silicon metal, 1·25% of manganese as ferromanganese, 1·3% of phosphorus as ferrophosphorus (20%), and, when molten, 0·3% sulphur was added as roll sulphur. One ingot was cast from this melt, and its structure showed very coarse flake graphite with islands of nickel phosphide (Fig. 51).

Other melts were carried out in an attempt to cause the formation of flake graphite by inoculation with ferrosilicon and electrode carbon, but no success was achieved. Attempts to remelt ingots of nickel–carbon alloys resulted in a very heavy loss of carbon, so that no eutectic graphite was visible in the resulting ingots. Similarly, attempts to desulphurize nickel–carbon melts containing sulphur by treatment with sodium carbonate caused an almost complete loss of carbon, giving ingots consisting of polyhedral grains surrounded by nickel sulphide (Fig. 74). It would appear that the carbon in nickel–carbon melts is very reactive to atmospheric oxygen and, probably, carbon dioxide.

A summary of the above melts and the inferences to be drawn from them is given in Table I.

It seems possible to draw the following tentative conclusions from this series of nickel–carbon melts :

(1) Undercooled, flake, and nodular structures, closely resembling those found in cast irons, can be reproduced at will.

(2) Additions of calcium silicide to the melt cause the formation of nodular structures.

(3) Additions of tellurium or sulphur give coarse flake graphite in all nickel–carbon alloys.

(4) In respect of sulphur additions, nickel–carbon alloys behave similarly to iron–carbon–silicon alloys.

(5) The coarsening effect of sulphur persists whether present as Ni_3S_2 or MnS.

(6) Calcium silicide additions in the presence of Ni_3S_2 or MnS have little or no effect.

(7) Treatment with hydrogen may coarsen the graphite.

(8) Treatment with hydrogen after treatment with calcium silicide prevents the formation of nodules.

(9) Treatment with titanium and carbon dioxide tends to coarsen the graphite.

(10) Treatment with titanium and carbon dioxide after treatment with calcium silicide prevents the formation of nodules.

(11) Kish graphite may inoculate the melt.

(12) Undercooling may occur.

Preliminary Cooling-Curve Experiments and Effect of Cooling Rate

It was felt that in the preliminary experiments on nickel–carbon alloys very little control had been exercised over cooling rates by casting ingots in green-sand moulds and that, in addition, variations in pouring temperature were certain to have occurred. Therefore the effect of cooling rate was investigated using a different melting technique. In each case 60 g. of pure nickel shot were melted in a graphite crucible machined from unused electrode carbon. The internal dimensions of each crucible were approximately 1 in. dia. by $3\frac{1}{4}$ in. deep, and the wall thickness was about $\frac{1}{4}$ in. A thermocouple in a thin silica sheath was placed through a hole in the graphite lid so that the hot junction was at the centre of the ingot. The thermocouples for all the work reported here were Johnson-Matthey platinum/platinum–10% rhodium, 0·25 mm. in dia., and they were sheathed with long sealed-end translucent silica tubing 0·125 in. in dia. The crucible was placed in each case at the centre of a platinum-wound tube furnace $2\frac{1}{8}$ in. in bore and 20 in. long, and was supported on a graphite stand, from which it was insulated by a $\frac{1}{4}$-in. thick silica disc. The charge absorbed carbon from the crucible walls to the maximum amount soluble at the highest temperature reached, that is, 1500° C. in each case. From this treatment every ingot became hypereutectic.

To investigate the effect of cooling rate, four ingots were made ; these were cooled as follows :

Ingot No.	Cooling
15	From 1500 to 1100° C., at rate of 100° C./20 min.
16	From 1500 to 1100° C., at rate of 100° C./12 min.
17	From 1500 to 1100° C., at rate of 100° C./8 min.
18	Taken from furnace at 1500° C. and air-cooled.

Ingots 15, 16, and 17 all had similar structures. The central zones in each case were entirely devoid of kish graphite, appearing exactly

eutectic in composition ; a typical spot is illustrated in Fig. 53. With decreasing cooling rate the graphite in these zones becomes slightly coarser. At the tops of the ingots segregations of kish graphite occurred ; a typical area is illustrated in Fig. 52. The amount of this segregation decreased with increasing rate of cooling. The flake graphite also became slightly coarser near the crucible walls and in the vicinity of the silica sheath of the thermocouple ; Fig. 54 shows the coarser graphite near the silica sheath of ingot 16. These melts are apparently susceptible to inoculation by graphite and to coarsening in the presence of silica. The latter point will be referred to later in more detail.

Ingot 18 had a structure completely different from the others cooled more slowly. At the centre of the ingot was a small area of undercooled graphite with primary metal dendrites and some distorted hypereutectic graphite ; this area is shown in Fig. 55. The remainder of the ingot had nodular graphite in the metallic matrix, with occasional distorted hypereutectic graphite flakes (Fig. 56). At a high magnification the nodules were found to be rather irregularly formed spherulites ; a typical example is shown in Fig. 57. Apart from the irregular outline, these graphite spherulites are exactly similar to those found in malleable iron of high sulphur content. We have now, therefore, two methods of producing nodular graphite structures in nickel–carbon alloys, *viz.*, (*a*) by treatment with calcium silicide or (*b*) by rapid cooling.

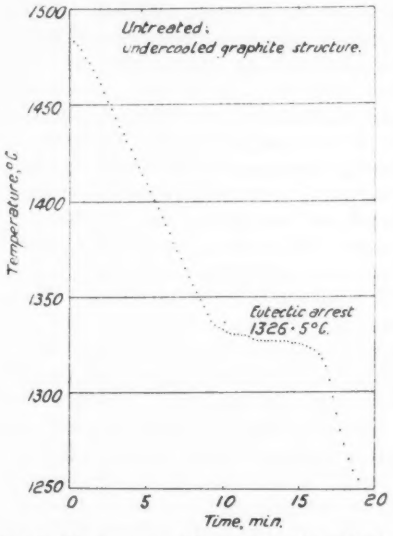


FIG. 127—Cooling curve of hypereutectic nickel–carbon alloy melted in a graphite crucible. Ingot 19. Superheated to 1500° C. No addition

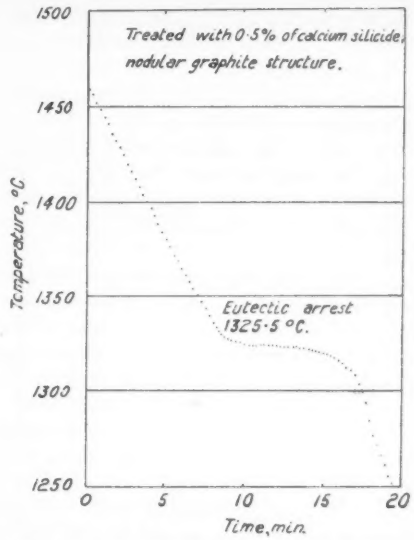


FIG. 128—Cooling curve of hypereutectic nickel–carbon alloy melted in a graphite crucible. Ingot 20. Superheated to 1500° C. and 0·5% of calcium silicide added to give a nodular graphite structure. (*See* Figs. 59 and 63)

Cooling curves were then taken on three further ingots melted in a manner similar to that described for ingots 15 to 18, except that in the case of ingots 20 and 21 a graphite tube was fitted to the top of the crucible leading out of the furnace. This facilitated the making of the additions. The temperature readings were taken on a null deflection potentiometer to 5 microvolts. The ingots were treated as follows :

Ingot No.	Treatment.
19	Superheated to 1500° C. No addition.
20	Superheated to 1500° C. Calcium silicide (0·5%) added.
21	Superheated to 1500° C. Sulphur (0·3%), as roll sulphur, added.

The additions were stirred in with a fine silica tube. The cooling curves obtained are given in Figs. 127, 128, and 129.

Many other cooling curves have been taken on nickel–carbon alloys in the untreated condition and after treatment with calcium silicide and sulphur. In all these the eutectic arrest for undercooled graphite and for spherulitic structures occurred at almost identical temperatures—the actual values all lay between 1323° and 1325° C. In the case of the coarse graphite structures produced by the addition of sulphur, the eutectic arrest occurred at temperatures ranging between 1310 and 1330° C., the actual temperature depending upon the sulphur content. No

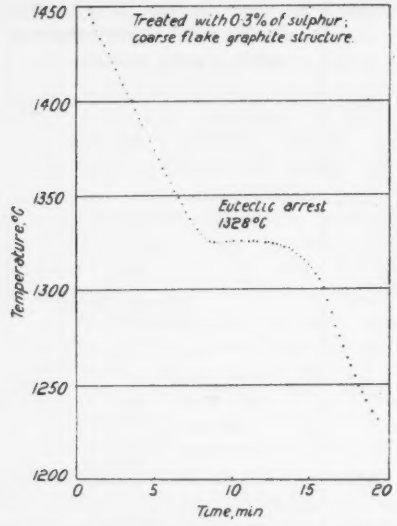


FIG. 129—Cooling curve of hypereutectic nickel-carbon alloy melted in a graphite crucible. Ingot 21. Superheated to 1500° C. and 0·3% of roll sulphur added to give coarse flake graphite. (*See* Figs. 60 to 62)

information was available upon the nickel–carbon–sulphur system, but the nickel–sulphur system[47] shows a sharp drop in the liquidus temperature with increasing sulphur contents until a eutectic of nickel and nickel sulphide occurs at 645° C. and 21·5% of sulphur. From this it might be expected that the addition of sulphur to nickel–carbon alloys would cause a drop in the temperature of the eutectic arrest. If this is true, it would explain the coarse graphite eutectic solidifying at various temperatures which may be lower than in those of untreated alloys with undercooled structures. In the case of the cobalt–carbon alloys subsequently to be described, it was possible to show that a coarse flake graphite alloy had a much higher eutectic arrest than an alloy of the same sulphur content with undercooled graphite. No method was discovered to produce coarse flake graphite in nickel–carbon alloys without the addition of either sulphur or tellurium, and it is possible that the latter element has a similar effect to that of sulphur in lowering the eutectic arrest temperature, irrespective of the type of graphite structure.

The microstructure obtained in ingot 19 was very similar to that described and illustrated for ingots 15 to 17. Ingot 20 had well-developed spherulites over most of its length* (Fig. 58), except near the top, where a very heavy segregation of spherulites occurred (Fig. 59). It would seem that these spherulites represent the hypereutectic graphite. If so, the nodules must have formed either directly from the melt or from another solid phase after the eutectic had solidified. Only by such means could the upward flotation be explained. The structure of a well-formed spherulite at a high magnification is shown in Fig. 63. The structure of ingot 21, treated with sulphur, had uniformly coarse graphite (Fig. 62), becoming very coarse and in greater amount at the top of the ingot. The graphite in this sample, being coarse, was easy to examine for internal structure, and under polarized light the flakes were found to have similar structures to those found in the graphite flakes of cast irons and reported elsewhere.[10] Typical examples of the internal graphite structures are given in Figs. 60 and 61.

* For micro-examination, complete vertical sections were taken of each ingot.

OCCURRENCE OF NODULAR STRUCTURES IN "AS-CAST" CAST IRONS.

Thus far it had been possible to show that nickel–carbon alloys possess many structural features in common with cast irons, but they appeared at first to differ in one unique and very important aspect, *viz.*, that nodular graphite structures of the spherulite type could be produced easily whenever desired in as-cast materials without any heat-treatment. In cast irons, nodular structures could be obtained only by the heat-treatment of a white cast iron at an elevated temperature for a lengthy period. It is of great commercial importance that when the graphite is in this form the cast irons possess considerable ductility and shock resistance, whereas with the graphite in the flake form they are non-ductile and brittle. If nodular structures of graphite could be produced in cast irons without heat-treatment, the material would assume, by virtue of its low cost and enhanced properties, much greater competitive value. With these considerations in mind it was decided to examine a wide range of cast irons to find whether nodular structures did occur in the as-cast condition.

Several instances of nodular structures in as-cast cast irons were found, some apparently forming by different mechanisms. Nodular structures were found associated with inverse chill; the following instance will suffice: Reference has been made to the ease with which inverse chill occurs in refined graphite irons having a completely undercooled graphite structure, and to the fact that complete refinement of the graphite can be achieved by additions of ferrosilicon-

zirconium. A charge of Swedish white iron and steel scrap was melted in a coke-fired forced-draught crucible furnace; when molten, ferro-silicon-zirconium was added to give material of the following analysis after pouring into 1·2-in. test-bar moulds in green sand:

Total carbon... ...	3·28%
Silicon	1·19%
Manganese	0·53%
Sulphur	0·023%
Phosphorus	0·030%
Zirconium	0·547%

The general structures of these bars is shown in Fig. 64 and consisted of fine undercooled graphite in a matrix of pearlite. The usual complex zirconium sulphide and zirconium carbide particles were present in this material. An area of inverse chill is illustrated in Fig. 65. Nodules of graphite are seen in transformed austenite dendrites with the interdendritic regions containing a laminated or "platey" white-iron eutectic. Fig. 66 shows the structure of this white-iron eutectic at a high magnification, where its pattern is clearly registered. In the eutectic of a normal white-iron structure the austenite or transformed austenite is the dispersed phase and the cementite is the continuous phase, but in this case the cementite phase shows little approach to continuity. Fig. 67 shows a nodule at a higher magnification. The peculiar nature of the white-iron eutectic in these illustrations should be noted and compared with that usually occurring in white irons (many examples are shown in the literature); reference will be made to this in a later section. Without doubt here we have a clear occurrence of spherulitic graphite nodules in cast iron which has not been subjected to heat-treatment.

A further case of the occurrence of nodules was found in a pair of cast-iron scissors of very thin section ($\frac{1}{8}$ in.); the micrograph of the structure (Fig. 68) shows a ferrite matrix with a network of phosphide eutectic and pearlite. The composition of the iron was:

T.C.	Si	Mn	S	P
3·3%	3·4%	0·66%	0·120%	1·43%

Taking into account both silicon and phosphorus, it will be observed that the iron is hypereutectic. Hanemann and Schrader [76] have illustrated a case of the occurrence of graphite spherulites in a single-cast piston ring, 80 mm. in dia. and 3 × 3 mm. in section, of the following composition:

T.C.	Si	Mn	S	P
3·75%	2·59%	0·50%	0·088%	0·54%

Again the iron was hypereutectic and cast in a very thin section. Several cases of this type have now been observed in hypereutectic cast irons which have been rapidly cooled. Occasionally the spherulites, which are usually very small, have flake graphite attached, as shown in Fig. 69. At this point it is convenient to re-examine the structure illustrated in Fig. 25. This, it will be recalled, shows the central zones of an ingot melted *in vacuo*, using a graphite crucible. The hypereutectic graphite has floated to the top of the ingot and in the central zones undercooling has occurred, giving transformed austenite dendrites and undercooled graphite. Fine nodular graphite is also to be seen in this micrograph. Nipper [8] has shown a similar case of spherulitic graphite in a vacuum-melted hypereutectic iron, and Gröber and Hanemann [1] have illustrated graphite spherulites in a sample containing 7·7% of carbon, quenched from 2250° to 160° C. It would thus appear that spherulitic carbon nodules can be produced by fairly rapid cooling of hypereutectic melts. At this stage no evidence was available to indicate whether these nodules formed directly from the melt or by the decomposition of a carbide after solidification. In some of the cases quoted, undercooling had resulted from rapid cooling, but in other cases undercooling was produced by vacuum melting, and the cooling rates were slow. To produce spherulites of this type, therefore, it seems that severe undercooling of a hypereutectic cast iron must occur. It is conceivable that these nodules represent the hypereutectic constituent, which may be graphite or carbide, deposited down the supersolubility curve of the hypereutectic side of the constitutional diagram (Fig. 7*c*). This would not, however, explain the presence of primary austenite dendrites. A further explanation would seem to be that undercooling follows the extension of the normal hypereutectic liquidus until the hypo-eutectic side of the supersolubility curve is reached. The graphite deposited over this temperature range would correspond to the spherulitic variety. The melt would then deposit primary dendrites of austenite until the hypertectic point is reached, when a binary complex of austenite and cementite solidifies and subsequently decomposes to give undercooled graphite (Fig. 7*d*). However, in the authors' opinion, a more likely explanation would be that solidification proceeds along the normal hypereutectic liquidus, depositing normal kish graphite, until the eutectic point is reached. From this point solidification then follows the undercooled extension of the normal hypo-eutectic liquid giving austenite dendrites. Further solidification follows the supersolubility curve on the hypereutectic side until the hypertectic point is reached. The

spherulitic graphite originates from the phase deposited down this portion of the supersolubility curve (Fig. 7*e*).

If these spherulites form as a result of the decomposition of cementite deposited down the hypereutectic supersolubility curve then we have an interesting extension of the findings of Eash that the undercooled eutectic solidifies first as a white-iron eutectic. The mechanism of spherulite formation in nickel–carbon alloys was unknown, and spherulite formation in heat-treated malleable iron was the only case where definite knowledge of the process existed. Could spherulites be formed directly from the melt ? A search was therefore made through the literature to find whether any general rules existed about undercooling and the formation of spherulites. The authors were impressed by the general lack of information on spherulitic crystallization. References 77 to 91 were among the most important examined. The undercooling of a few organic- and inorganic-salt eutectics has been studied in some detail, but little or no reference is made to the type of crystallization produced by different degrees of undercooling. Spherulitic formations are well known to the mineralogist, but practically no information is available from this source upon how they form. One investigator briefly refers to the production of spherulites by the solidification of liquid crystals and also by the rapid cooling of a molten organic compound into the region of labile crystallization. This latter point is particularly interesting in view of the fact that rapid cooling of nickel–carbon alloys gives graphite spherulites. Two American investigators have produced artificial spherulites and related aggregates by allowing a concentrated solution of one salt to diffuse into a gel containing a dilute solution of another salt able to react with the first by metathesis. It should be noted that these conditions do distantly resemble those obtaining when spherulites are formed by heat-treatment of a white iron (containing iron sulphide). These investigators point out that spherulites must grow by the simultaneous crystallization of fibres radiating from a common centre in all directions with equal speed. Such a mechanism would lead to a spherical shape at all times during their growth. However, during the course of their experimental work and in agreement with certain German investigators, they advance the hypothesis that spherulitic crystallization may also begin from a centre within two opposite solid angles, together less than 360°, and that at a further stage the outward growth of the fibres ceases and further crystallization takes place by filling-in the waist. No evidence of this mode of spherulitic formation has been found in any of the materials mentioned in this report.

Since it was well known that spherulites could be produced in cast iron by the decomposition of cementite during heat-treatment of a white iron at 650–1050° C., and as the possibility of producing graphite in this form directly from the liquid seemed too nebulous at this stage, an attempt was made — avoiding heat-treatment — to obtain spherulites, basing the procedure on known facts. The following gives the train of reasoning leading up to the experimental work in this context.

It was assumed that a mechanism was required which would enable an iron to solidify white and subsequently, after complete solidification, to graphitize without heat-treatment. This process would be essentially the same as that employed in the malleable-iron process. The only situation known definitely to lead to suitable conditions would have to be based upon the reaction between manganese and sulphur in cast iron. In the absence of manganese the sulphur exists as iron sulphide, but manganese (if present), having a greater affinity for sulphur, replaces it with manganese sulphide if added in sufficient excess. Iron sulphide causes the stabilization of iron carbide, whilst manganese sulphide behaves neutrally in this respect. The reaction between iron sulphide and manganese can be expressed as follows :

$$FeS + Mn \rightleftharpoons MnS + Fe.$$

This reaction is reversible, tending to proceed to the left-hand side at high temperatures and to the right-hand side at lower temperatures. Owing to this reversibility it is necessary, if no iron sulphide is required, to have manganese present in excess of that required theoretically to give manganese sulphide according to the formula MnS. Even when at room temperature only manganese sulphide is present, appreciable amounts of iron sulphide may be formed at elevated temperatures (1000–1600° C.), owing to the reversibility of the reaction. It was thought, bearing this in mind, that by adjusting the sulphur and manganese contents so that during solidification a sufficient amount of iron sulphide was present to cause white-iron solidification, and that if after solidification—consequent upon decreasing temperature—sufficient sulphur became converted to manganese sulphide, the white-iron structure would be enabled to graphitize automatically. Trials of a wide range of manganese and sulphur contents were made, but it seemed that for the principle to work the presence of an additional graphitizing element was required. With irons of the composition indicated below it was found possible to produce a form of graphite which

probably formed by the decomposition of cementite after solidification :

T.C., %	S, %	Mn, %	S, %	Ni, %
2·8–3·2	0·7–1·3	0·1–0·3	0·2%–0·6	1·5–7·0*

* Or 1–6% of copper.

The type of graphitization produced by such compositions has been described and illustrated in another paper,[92] dealing with the neutralization of sulphur by various alloying elements. It is generally characterized by a mesh-like pattern with no individual graphite flakes, all the graphite in one eutectic cell being interconnected. From the arrangement of some of this graphite it appears to outline what were originally islands of cementite and, occasionally, where graphitization is not complete, it can actually be seen rimming areas of eutectic carbide. On rare occasions a few very fine spherulites were observed. These have also been illustrated elsewhere.

It was felt that these findings justified the assumption that the manganese–sulphur reaction could be used to produce graphitization after solidification. This process could under certain conditions proceed to completion but, except on a very small scale, only a few isolated spherulites were produced. It was therefore decided that it would be necessary to provide a number of nuclei at which spherulitic crystallization could begin. After many experiments the following procedure was found to be slightly more successful in producing spherulites. Melts were made with compositions within the following range :

Total carbon... ...	2·6–3·2%
Silicon	1·8–2·6%
Manganese	0·1–0·3%
Sulphur	0·01–0·1%
Phosphorus	0·03–0·1%

When molten, the melts were treated with 1–3% of calcium silicide and then with about 0·1–0·4% of sulphur. It was argued that the calcium silicide would provide the nuclei and then the sulphur addition would provide the required manganese–sulphur reaction. No irons were produced by this method with all the graphite in the form of spherulites, although these were much larger and more numerous than previously ; typical examples are shown in Fig. 70. Occasionally the graphite assumed a pseudo-Widmanstätten pattern (Figs. 71 and 72), and sometimes occurred in a very fine spherulitic pattern attached to flakes of graphite (Fig. 73).

We have now described three distinctly different occurrences of spherulitic graphite formations in as-cast cast irons, *viz.* :

(1) Associated with inverse chill in irons also containing undercooled graphite.

(2) In undercooled hypereutectic grey irons.

(3) In irons of low manganese and high sulphur contents.

With this knowledge there appeared to be justification for proceeding with the investigation of nickel–carbon spherulitic formations, but, before describing this, one other group of experiments must be referred to. Remembering that additions of calcium silicide to nickel–carbon alloys did not produce spherulites in the presence of sulphur, it was thought that if this element could be eliminated in cast irons, spherulites might also be produced by calcium silicide. Synthetic melts were made with sulphur contents as low as 0·001%, but all these reacted in the normal way with calcium silicide, giving flake graphite.

The Effect of Calcium Silicide on Nickel–Iron–Carbon Alloys

As calcium silicide did not produce graphite spherulites even in cast irons of very low sulphur content, it was felt that the production of spherulites by this method may be a property of nickel–carbon alloys alone. It was therefore decided to investigate the effect of iron upon this tendency in a series of hypo-eutectic nickel–iron–carbon alloys ranging from the nickel–carbon to the iron–carbon side of the ternary system. The ternary iron–nickel–carbon system has been investigated by Kase[70] and his results have been redrawn and criticized by Marsh,[93] who has also included those obtained by Schichtel and Piwowarsky.[94] Many points are still in doubt, but with safety it can be said that the liquidus surface shows a trough extending from the binary eutectic on the iron–carbon to the binary eutectic on the nickel–carbon side. Fig. 130, taken from Marsh, indicates the effect of nickel on the carbon content and temperature of the eutectic.

Eleven melts were carried out in a high-frequency induction furnace, using charges made up of pure nickel, electrode carbon, and Lowmoor iron. A representative analysis of the iron was :

C	Si	Mn	S	P
0·18%	0·02%	0·14%	0·004%	0·026%

After superheating to about 1500° C., one ingot 6 in. long by $\frac{1}{2}$ in. dia. was poured in green sand from every melt. About 0·8% of calcium silicide was then added and a further similar ingot was cast. Samples for chemical analysis and microscopic examination were taken from the bottom half of each ingot. The analyses and microstructures of these ingots are summarized in Table II. All the analyses were carried out on the calcium silicide treated bars.

Melts 22 to 29 were carried out first and then followed by melts 30 to 32, to scrutinize the

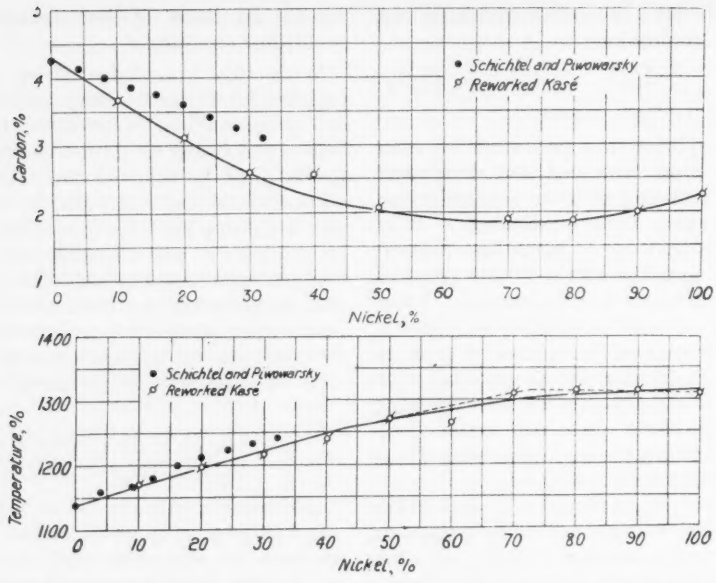


FIG. 130—Effect of nickel on the temperature and carbon content of the eutectic in iron–nickel–carbon alloys (Marsh)

TABLE II—*Effect of Calcium Silicide Additions to Nickel–Iron–Carbon Alloys*

Melt No.	Fe, %	Ni, %	C, %	S, %	Si, %	Treatment	Microstructure
22	5·74	...	1·60	...	...	Nil	Undercooled Nodular
						CaSi	
23	9·84	...	1·50	...	...	Nil	Undercooled Nodular
						CaSi	
24	23·8	...	1·72	...	...	Nil	Undercooled Nodular
						CaSi	
25	40·5	...	1·57	...	...	Nil	Undercooled Nodular
						CaSi	
26	57·86	...	1·57	...	...	Nil	Undercooled Flake
						CaSi	
27	74·33	...	2·22	...	...	Nil	Undercooled Flake
						CaSi	
28	85·79	...	2·41	...	...	Nil	Undercooled Flake
						CaSi	
29	90·81	...	2·87	...	...	Nil	Undercooled Flake
						CaSi	
30	46·83	50·72	1·58	0·01	0·89	Nil	Undercooled Nodular
						CaSi	
31	48·79	48·06	2·08	0·01	1·00	Nil	Undercooled Nodular
						CaSi	
32	53·0	44·61	1·93	0·01	0·61	Nil	Undercooled Flake
						CaSi	

region 45–55% of iron in greater detail. Inspection of the results clearly reveals a change in the reaction of these alloys to additions of calcium silicide at some iron content lying between 49 and 53%. Typical microstructures of this series are represented by Figs. 75 and 76, taken from melt 31, and Figs. 77 and 78, taken from melt 32. From these results the present authors concluded that iron in amounts greater than about 50% confers, either directly or indirectly, upon nickel–iron–carbon alloys a property by virtue of which they are susceptible to nucleation by calcium silicide to give flake graphite. For iron contents of less than about 50%, calcium silicide behaves in a different manner and certainly does not inoculate the melt, spherulitic crystallization being produced.

Formation of Undercooled Graphite in Cast Irons

It will be recalled that reference was made to the work of Eash, who found that irons giving undercooled graphite structures solidify white and subsequently decompose to give undercooled graphite after solidification. It was also pointed out that this investigator's methods of demonstrating this might be considered somewhat crude and the composition of his irons unusual. In view of the importance of Eash's suggestions, the present investigators decided to try to confirm them, using different experimental methods. It should be stated immediately that the results which were eventually obtained entirely confirmed the claims made by Eash.

For these experiments two irons were used, one a low-carbon iron (in this respect similar to that used by Eash) and the other one of a more normal composition. The analyses of these two irons are as follows :

	Low-Carbon Iron	Normal-Carbon Iron
Total carbon, % ...	1·92	3·10
Silicon, % ...	2·68	2·05
Manganese, % ...	0·74	0·90
Sulphur, % ...	0·036	0·045
Phosphorus, % ...	0·03	0·04
Titanium, % ...	0·16	0·21
Vanadium, % ...	0·27	0·31

In contrast to the iron used by Eash, these irons contained titanium (and vanadium) in sufficient amounts to guarantee the formation of undercooled graphite structures, even with relatively slow rates of cooling.

Small ingots of these irons weighing about 40–50 g. were melted in fireclay crucibles in a platinum-resistance tube furnace placed vertically above a bath of water about 2 ft. deep. By moving a lever the bottom of the furnace could be instantly disengaged and the contents of the furnace, including the ingot in the crucible, plunged into the bath of water. Temperatures were measured using platinum/platinum–10% rhodium thermocouples and a null-deflection quick-reading potentiometer, readings being taken to 5 microvolts. Cooling curves were taken by recording the temperature every 15 sec.

Melts of the high-carbon material were made first. It was found that ingots slowly cooled to room temperature in the furnace gave an undercooled graphite structure such as that illustrated in Fig. 79 of the ingot from melt 33. On many occasions ingots of 40–50-g. weight were quenched within 5° C. below the eutectic arrest. In no case was a completely white structure produced, but in many a partially white ingot was obtained. Fig. 131 shows the cooling curve of ingot 34, which was rapidly quenched immediately after solidification. Fig. 80 shows a vertical section through this ingot, and Fig. 81 shows the white-iron structure at a higher magnification. The grey portions of the melt had undercooled graphite in a matrix of martensite. Similar experiences were obtained with the low-carbon material ; by quenching immediately after solidification no completely white samples were obtained.

Whilst the delay between the taking of the last temperature reading and plunging the ingot into the quenching bath was only a fraction of a second, it was felt that either the effectiveness of the quench was inadequate, owing to the crucible surrounding the ingot, or that the breakdown of

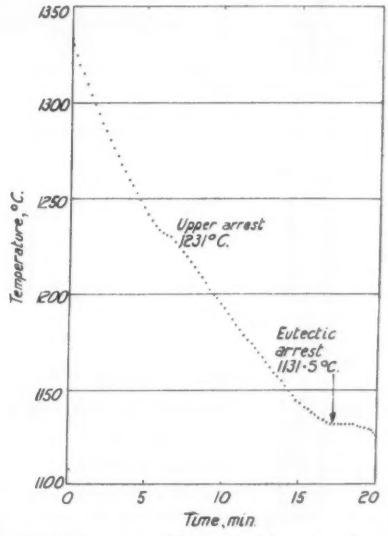


Fig. 131—Cooling curve of 50-g. cast-iron ingot (ingot 34) quenched immediately after solidification. (*See* Figs. 80 and 81)

the undercooled white iron was very much more rapid than was indicated by the work of Eash. It was further possible that the cooling curves were not indicating complete solidification. Accordingly, it was decided to use very much smaller ingots, *viz.*, of from 5 to 7 g. These were in crucibles made from silica tubing, and the arrangement, together with the approximate dimensions, is illustrated in Fig. 132. Using this arrangement ingots were obtained which were almost completely white when quenched just after solidification. A typical cooling curve for a melt of the higher-carbon material, using a 6-g. ingot, is shown in Fig. 133, and the structure of a section of the quenched ingot is seen in Fig. 84. The duration of the arrest in this case was as long as that of similar ingots which were allowed to cool to room temperature at the same initial rate, but which were completely grey. The very small grey portions visible in this ingot could not have been responsible for the arrest.

From these results it was apparent that the breakdown of this carbide in undercooled samples is extremely rapid and may even be proceeding before the whole of the eutectic is solid. It should be noted that the eutectic found in these quenched samples (Figs. 81 and 84) is very similar to that found in the inverse chill of fine undercooled graphite irons (Figs. 65 and 66). The

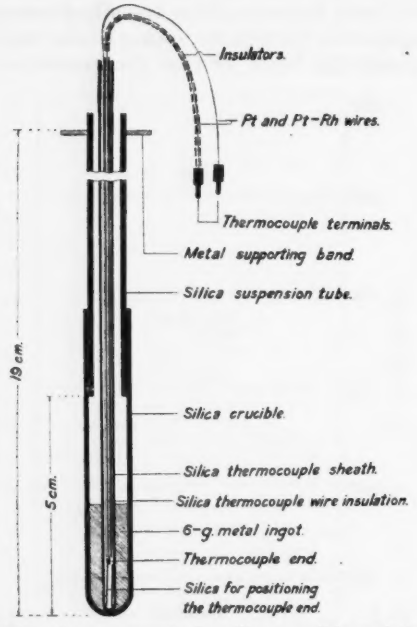


FIG. 132—Arrangement of a silica crucible and thermocouple for quenching experiments with 5–7-g. ingots

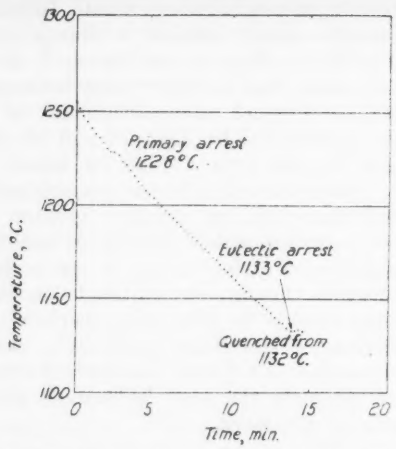


FIG. 133—Typical cooling curve obtained with a 6-g. cast-iron ingot melted in the arrangement shown in Fig. 132. (For structure *see* Fig. 84)

same plate-like characteristics are visible. That this structure differs from that of a normal white iron is shown by Fig. 82, which illustrates the general structure of a 50-g. ingot of the higher-carbon material to which an addition of 2% of chromium had been made and which was cooled at the same rate, during solidification, as the ingot illustrated in Fig. 81. This may provide a clue to the formation of inverse chill in undercooled irons, since the white iron represents the eutectic, which, in that particular locality, instead of decomposing after solidification to give undercooled graphite, remains stable to room temperature. Occasionally the carbide of the inverse chill does begin to graphitize at a lower temperature than is required to give undercooled graphite, giving nodular structures of the type shown in Figs. 65, 66, and 67. Why the white-iron structure at the centre of the section should show this increased persistence is not apparent, but it should be remembered that it shows other features of metastability differing from normal white iron, for instance, its apparent immunity to graphitizing elements, etc.

Formation of Undercooled Graphite in Nickel–Carbon Alloys

As undercooled graphite structures, as well as flake graphite, could be produced in nickel–carbon alloys, it was decided to investigate whether in these alloys also an unstable eutectic solidified first and then decomposed to give undercooled graphite. Initially the experiments were carried out on 50-g. ingots of hypo-eutectic melts in fireclay crucibles, but, as with the cast irons, no

positive results were obtained. Next, 6-g. ingots were melted in silica crucibles similar to those used for cast irons, but the nickel–carbon melt reacted with the silica with a marked evolution of gas, which invariably lifted the melt to the top of the crucible and even blew the molten metal up the very narrow space between the thermocouple sheath and the silica positioning tube. This probably reflects an important difference between cast irons and nickel–carbon alloys, the carbon in the latter being much more reactive to oxides and oxygen. This has been referred to previously in connection with the remelting and desulphurizing with Na_2CO_3 of nickel–carbon alloys. In the particular case referred to here, the following reaction probably occurs :

$$2C\ (Ni) + SiO_2 = Si + 2CO$$

The experiments were therefore continued using 6-g. ingots melted in graphite crucibles of the same size as the silica crucibles used previously. This method proved satisfactory. Sharp arrests were obtained, and, owing to the high thermal conductivity of graphite, more efficient quenching resulted. Ingots were heated to 1400–1500° C. and cooled in the furnace. Those cooled to room temperature or quenched more than 50° C. below the eutectic arrest showed undercooled graphite with some kish graphite, particularly at the tops of the ingots. The crucible walls had the usual faint coarsening effect on the graphite. The ingots cooled in the same manner but quenched immediately after solidification had structures consisting almost completely of spherulitic graphite with, adjacent to the crucible wall, a thin band of flake graphite followed by a still thinner band of very fine undercooled graphite. Fig. 134 shows a cooling curve of a 6-g. nickel–carbon ingot up to the point of quenching, and Fig. 83 illustrates the general and edge structure of the same ingot. The coarse graphite produced by the crucible wall is clearly seen. The graphite nodules in this sample were spherulites exactly similar to those illustrated previously in this paper, but of much smaller size. It would appear from this group of experiments that undercooled graphite structures in nickel–carbon alloys arise as the result of the decomposition of an unstable eutectic, probably nickel solid solution + nickel carbide (Ni_3C), shortly after solidification. This carbide may be very much more unstable than that formed under similar conditions in cast irons and so decomposes at a somewhat lower temperature during the quench to give the spherulitic graphite which is seen in Fig. 83. Near the crucible wall the graphite crucible has inoculated the melt to prevent undercooling, and this flake graphite has in turn rendered the undercooled carbide bordering it more unstable than that some distance away so that it decomposed at its normal temperature, giving characteristic undercooled graphite.

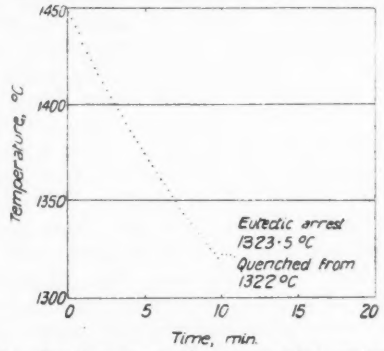


FIG. 134—Cooling curve of a 6-g. nickel–carbon ingot melted in a graphite crucible and quenched immediately after solidification. (For structure, *see* Fig. 83)

Now we have a clue to the formation of nodular structures in nickel–carbon alloys by rapid cooling (referred to previously, *see* Figs. 56 and 57). In cast irons which normally would give undercooled graphite structures, an increase in the rate of cooling gives progressively finer graphite until, in the limiting case, the cooling rate is so rapid that the white eutectic does not have time to transform and a white iron results. In nickel–carbon alloys a similar state of affairs exists, but although the very rapid cooling does not allow time for carbide decomposition at a high temperature, it cannot prevent it from decomposing at a lower temperature, owing to the extreme instability of the carbide. At lower temperatures spherulites are produced.

Formation of Spherulites in Nickel–Carbon Alloys by Calcium Silicide Additions

Examples of the effect of calcium silicide have already been shown and discussed, but the findings just described suggest an explanation of the calcium silicide effect in nickel–carbon and nickel-rich nickel–iron–carbon alloys. It will be recalled that the freezing point of the eutectic in calcium silicide treated ingots was very near to that of untreated ingots giving undercooled graphite. This has been confirmed several times, the temperature of the eutectic arrest in both cases being 1325° C. ± 1. It is believed that this temperature represents the solidification of the eutectic of nickel carbide and a solid solution of carbon in nickel. The function of calcium silicide is to stabilize the carbide over a greater range of

temperature, or to provide the right kind of nuclei for spherulitic crystallization, or a combination of these. The effect of the first possibility would be to cause carbide decomposition at a lower temperature than is normal. The fact that rapid cooling also produces spherulites in nickel–carbon alloys would seem to indicate that their production by the addition of calcium silicide is not necessarily a process involving the introduction of nuclei into the melt. It might be argued that cast irons differ in this respect from nickel–carbon alloys. This is true in so far as calcium silicide does not promote the stability of carbides in iron-rich alloys, but it is possible that the undercooled white iron, normally parent to undercooled graphite, may decompose at lower temperatures to give nodular structures ; this has already been quoted in the section preceding the previous one, as an explanation of spherulites occurring with inverse chill (*see* Figs. 65, 66, and 67).

If this is correct we now have the problem of producing spherulitic graphite in a cast iron without heat-treatment stated in a further form. The white-iron structure giving rise to undercooled graphite in cast irons already fulfils some of the conditions for nodule formation—it solidifies white and after solidification it is very unstable, changing to austenite and graphite. If this carbide can be stabilized to lower temperatures, or if the appropriate nuclei for spherulitic formation can be introduced, the problem will be solved.

So far in this discussion we have referred only to the spherulites produced in nickel–carbon alloys which originate from the eutectic portion of the alloy, but it has already been shown (Figs. 35 and 59) that hypereutectic graphite can also be obtained in this form by treatment with calcium silicide. Furthermore, very rapid cooling of hypereutectic melts will also give nodules. Thus if the explanations proffered so far are correct for the eutectic, they must also apply to the hypereutectic constituent. Now, in Figs. 58 and 59 it has been shown that the hypereutectic spherulites segregate at the tops of the ingots as if by flotation. It is suggested here that these most probably floated to that position—at least partially—as graphite, since nickel carbide would not be expected to have such a low relative density as to permit such a marked flotation. Therefore the hypereutectic graphite spherulites must have formed as the result of the decomposition of hypereutectic carbide suspended in the liquid. The process by which this occurs is not easy to imagine, because in the case illustrated the nodules are quite large and would therefore begin to rise in the melt before reaching their final dimensions and continue to grow on their journey upwards. For the graphitization of a hypereutectic nickel carbide during solidification, we have perhaps :

$$Ni_3C \longrightarrow 3Ni + C,$$

but in the case under consideration the reaction would take place in liquid surroundings, and the nickel product of the reaction would hold more than the eutectic concentration of carbon in solution and, furthermore, it would be liquid, since—according to the equilibrium diagram and the above considerations—the only phases which can be present in hypereutectic alloys above the eutectic temperature are liquid, graphite, or nickel carbide. Thus the explanation for the formation of spherulites by calcium silicide additions based upon an increased stability of a carbide becomes more difficult when applied to hypereutectic melts, and the possibility of calcium silicide providing the appropriate nuclei for spherulitic crystallization looms larger. The general literature on the subject of spherulitic formation indicates, without specifically saying so, that one condition for the process is restricted growth. In white cast irons being annealed, one of the present authors[96] has shown that the determining factor is whether the right nuclei (FeS) are present, and not the extent of the restriction of growth, since temperature has no effect on the type of crystallization.

Nickel Carbide in Nickel–Carbon Alloys

In what has been said so far, the existence of a nickel carbide has been tacitly assumed and it has been suggested that its existence in nickel–carbon alloys was not established. The compound had, however, been produced by the reaction of nickel powder and carbon monoxide. By quenching hypereutectic ingots in water directly from the liquid state nodular structures predominated, but occasionally areas of graphite occurred which were obviously of interdendritic pattern (Fig. 85). It is suggested that this might represent an area where carbide broke down *in situ* because the time necessary for diffusion to spherulites was not allowed. The graphite in these areas was very difficult to polish, seeming to be very loosely held in position. With calcium silicide treated ingots, quenching from the liquid produced many more of these areas with much greater ease. For instance, Fig. 86 shows the structure of a 200-g. hypereutectic ingot treated with calcium silicide and quenched about half-way through the eutectic arrest. Again, it should be noted, many well-developed nodules have formed and many dark, apparently interdendritic, areas of graphite can be seen. The increased ease with

which these are formed in calcium silicide treated ingots could be explained by the effect of this addition in increasing the stability of the carbide.

A number of 6-g. ingots were melted in graphite crucibles and held molten for lengthy periods. After quenching, these ingots did apparently show a nickel carbide. For instance, Fig. 87 shows the structure of an untreated ingot heated to 1500° C., slowly cooled over half an hour to 1450° C., and then quenched rapidly in a bath of water. Nodules of graphite are seen, together with a half-tone constituent. At high magnifications these areas can be resolved into an apparent Widmanstätten structure of nickel carbide in solid solution (Fig. 88). Evidence that the dispersed phase is a carbide is given in the proximity of graphite particles ; the phase always decreases in amount near to, and is never found adjoining, graphite (Fig. 89). Confirmatory evidence was obtained when a portion of the ingot was annealed at 700° C. for 4 hr., the dispersed phase in the Widmanstätten pattern being replaced by fine spots of graphite (Figs. 90 and 91).

Marsh [93] has commented upon Kase's [70] assumption of the complete intersolubility of iron and nickel carbide as a very improbable possibility. This criticism is presumably based upon the idea that nickel carbide [72] has a different structure from that of iron carbide.[97, 98] In view of the uncertainty which still exists about the structure of iron carbide it is perhaps not wise to go beyond this, but in support of Marsh's contention we have this apparent change in carbide stability and behaviour in the presence of calcium silicide, occurring somewhere in the region of 50% of iron in the nickel–iron–carbon alloys.

Effect of Silicon on Response of Nickel–Carbon Alloys to Calcium Silicide Additions

Four melts of pure nickel, together with 1%, 2%, 5%, and 10% of silicon respectively (as silicon metal), were carried out in graphite crucibles. The melts were slowly cooled in a platinum-wound furnace. This series was then repeated, but 0·5% of calcium silicide was added in each case before cooling. A summary of the microstructures is given in Table III.

Table III—*Effect of Silicon on Nodule Formation in Nickel–Carbon Alloys*

Silicon Added	Microstructure of Untreated Ingots	Microstructure of Ingots Treated with CaSi
1%	Undercooled graphite	Nodular graphite
2%	Undercooled graphite + traces of nodular graphite + a few particles of coarse graphite.	Nodular graphite
5%	Coarse graphite + nickel silicide (Ni_3Si_2) ...	Nodular graphite + nickel silicide
10%	No graphite visible. Solid solution + eutectic with Ni_3Si_2	No graphite visible. Solid solution + eutectic with Ni_3Si_2

Quenching experiments were also carried out on nickel–iron–carbon melts with iron contents of between 40 and 60%. Melts containing more than about 50% of iron had eutectic carbides after quenching immediately on completion of eutectic freezing. Nodular structures were also present in these alloys. Melts containing less than about 50% of iron and correspondingly more nickel gave no carbides when quenched in the same manner. There seems to be a change in carbide stability at about 50% of iron in these alloys, which roughly corresponds with the point at which they react differently to calcium silicide. The structure of an alloy with approximately 60% of iron, quenched immediately after solidification, is seen in Fig. 92, and Fig. 93 shows the carbide phase at a higher magnification. Fig. 94 shows the structure of an alloy with approximately 40% of iron quenched immediately after solidification. No carbides are visible in this case ; only fine graphite nodules and a small amount of interdendritic graphite are to be seen.

In amounts of up to at least 5%, silicon does not appear to influence the tendency of calcium silicide to give nodular graphite when added to nickel–silicon–carbon alloys. The microstructure of the 5% silicon ingot treated with calcium silicide is shown in Fig. 95, in which nodular graphite and nickel silicide in a solid-solution matrix can be seen. The phases introduced by the addition of silicon were similar in appearance to those shown by Iwaze and Okamoto [99] in the nickel–silicon alloys. At 10% of silicon no graphite was visible, and therefore it would appear that silicon either reduces the solubility of graphite or stabilizes the nickel carbide. Ingots containing 5% and 10% of silicon, respectively, were annealed for varying periods at 900° C., but no more graphite appeared nor did the amount of nickel silicide decrease, so that silicon probably reduces the amount of carbon dissolved.

Effect of Copper on Response of Nickel–Carbon Alloys to Calcium Silicide Additions

Nickel–copper–carbon alloys were prepared with up to 20% of copper. In the untreated condition undercooled graphite was obtained in all cases, and nodular graphite was obtained in every case after treatment with calcium silicide.

The Effect of Alloy Additions to Nickel–Carbon Alloys, the Alloy Additions Containing Calcium or other Alkaline-Earth Elements

Since additions of silicon did not have any effect on the graphite structure of nickel–carbon alloys, it was considered likely that the effect of calcium silicide was due to its own peculiar crystal structure functioning as an appropriate nucleus for graphite formation, or to the alloying influence of calcium. To investigate this a calcium–silicon alloy containing more silicon than the calcium silicide used previously,* and various other alloys containing calcium or other alkaline-earth elements,† were used. The compositions of these alloys are given in Table IV. According to the calcium–silicon equilibrium diagram[100] the calcium silicide used in these experiments so far should contain CaSi and $CaSi_2$. The calcium–silicon alloy listed in Table IV contained 80% of silicon, and hence should have consisted of silicon and $CaSi_2$.

Table IV—*Alloys Containing Alkaline-Earth Elements*

Alloy	Composition
Ca–Si	Si, 80%.
Ca–Mg	Ca, approx. 40%; Na, < 2%; Cl, not more than 0·5%; Mg, remainder
Ca–Cu	Ca, approx. 40%; Na, < 1%; Cl, not more than 2%; Cu, remainder
Ca–Zn	Ca, approx. 40%; Na, < 1%; Cl, < 1%; Zn, remainder
Ca–Pb	Ca, approx. 10%; Na, < 0·5%; Cl, < 0·5%; Pb, remainder
Sr–Mg	Sr, approx. 25%; Na, < 2%; Cl, < 2%; Mg, remainder
Ba–Mg	Ba, approx. 35%; Na, < 2%; Cl, < 2%; Mg, remainder
Ba–Pb	Ba, approx. 30%; Na, < 2%; Cl, < 1%; Pb, remainder

Ingots of 60-g. size were melted in graphite crucibles, using pure nickel shot. As previously, the crucibles were machined from electrode carbon. The melts were heated to about 1470° C. and were then treated with 0·5% of one of the alloy additions.

Well-formed spherulitic graphite was obtained in every instance, with the exception of the melt treated with barium–lead, when the graphite was in the form of fine flakes with some irregular nodules which were not spherulitic. This structure is illustrated in Fig. 96. In every ingot treated with an alloy containing magnesium the eutectic structure of $MgNi_2$ was seen.[101] Fig. 97, showing the structure of an ingot after treatment with calcium–magnesium, and Fig. 98, showing the $MgNi_2$ structure at a high magnification, illustrate this. Unless the addition was stirred into the melt, the magnesium-containing phases and the nodular structures occurred only at the top of the ingot, the remainder of the graphite being in the undercooled form. When stirred in, however, the magnesium-containing phases and the nodular structures were both uniformly distributed. This appears to relate the presence of alloyed magnesium with the formation of spherulites. Traces of undercooled graphite were also found in the ingots treated with calcium–zinc. Similarly, additions of calcium carbide have also been found to produce spherulitic graphite in nickel–carbon alloys.

Since these experiments indicated that magnesium alloyed with nickel–carbon alloys was a factor in the production of spherulites, additions of magnesium were made to a number of melts, using magnesium metal, a nickel–magnesium alloy (Mg, 30%), and a copper–magnesium alloy (Mg, 43%).‡ In every case graphite spherulites were produced. It can be concluded that magnesium and calcium most probably function as alloying elements in giving nodular structures. Barium and strontium may behave in a manner similar to calcium and magnesium, but when a positive effect was obtained from alloys containing them magnesium was also present.

Summarizing, it is felt that the essential feature of an addition for the production of spherulites in nickel–carbon alloys is that it shall contain at least either calcium or magnesium in some

* The analysis of the calcium silicide used in all the experiments described so far was : Silicon 62·09%; calcium 25·1%; aluminium 1·39%.

† These alloys were given for the investigation by the Imperial Chemical Industries, Ltd.

‡ Kindly given by Messrs. J. Stone & Co., Ltd.

suitably protected form, so that either element can enter the melt. No evidence has been brought to light to indicate that the crystal structure of the addition has any influence.

Experiments with Cobalt–Carbon Alloys

Much less work has been carried out on the cobalt–carbon alloys than upon the nickel–carbon alloys, but sufficient data have been accumulated to deepen the analogy between iron–carbon–silicon alloys on the one hand and nickel–carbon and cobalt–carbon alloys on the other. No experiments were carried out on the ternary alloys of cobalt and carbon with iron, silicon, and nickel, but these should provide an interesting field for future investigation.

The cobalt–carbon diagram has been studied by several investigators[102–104] whose work shows a fair measure of agreement in respect of the general features of the liquidus. The high-temperature regions in the system are very similar to those in the nickel–carbon system, showing a eutectic at 1300–1315° C., at 2·8–3·0% of carbon. There has been some disagreement as to the actual location of the eutectic point; Boecker[102] put the eutectic temperature at 1300° C., while Hashimoto[104] put it at 1315° C. The eutectic consists of a solid solution of carbon in cobalt and graphite. Cobalt does not appear to form a stable carbide, the only solid phases reported being graphite and solid solution.

Cobalt from several sources was used in these experiments, and some was discarded owing to the presence of impurities, notably sulphur. For the first experiments cobalt briquettes were used, and spectrographic examination of these revealed the following impurities, in order of amounts present: Silicon, manganese, nickel, magnesium, and iron. Chemical analysis gave 0·290% of sulphur in these briquettes.

By air-cooling such ingots the eutectic graphite was obtained in the fine undercooled form, the general structure showing primary graphite and primary solid solution typical of undercooled alloys. The addition of 0·3% of sulphur to air-cooled ingots gave fairly coarse flake graphite. Again, as in the iron–carbon–silicon and nickel–carbon alloys, we have the coarsening influence of sulphur.

Hypo-eutectic melts made up with the above material and electrode carbon, melted in fireclay crucibles and then air-cooled, gave mixed structures of flake and undercooled graphite such as are shown in Fig. 99. Examination of these alloys at a high magnification revealed the presence of cobalt sulphide (Fig. 100). A purer form of cobalt was then obtained in powdered

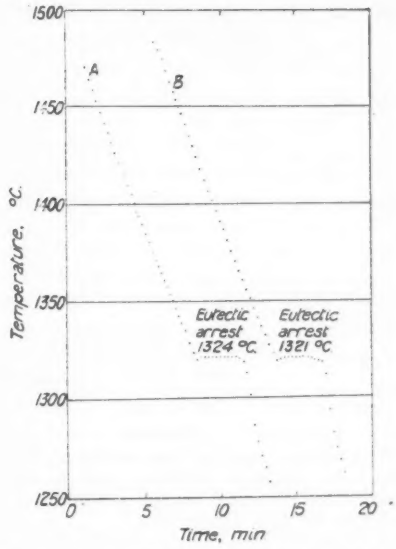


Fig. 135—Cooling curves of two 60-g. cobalt–carbon ingots melted in graphite crucibles: *A*, treated with calcium silicide; *B*, untreated. (For structures *see* Figs. 101 and 102)

form. Chemical analysis showed the following impurities:

Ni	Mn	S	CaO	Al_2O_3	C
0·05%	<0·10%	0·033%	0·99%	0·045%	0·017%

Fe	Cu	P	MgO	SiO_2
0·22%	0·15%	<0·01%	0·24%	0·20%

Two 60-g. melts were carried out using this material melted in graphite crucibles, and cooling curves were taken. The melts were heated to 1500° C. before cooling, one being treated with 0·5% of calcium silicide. The cooling curves obtained are given in Fig. 135. In the untreated sample the eutectic arrest occurred at 1321° C. and in the calcium silicide treated ingot it occurred at 1324° C. The central zones of the untreated ingot had fairly coarse graphite (Fig. 101), the calcium silicide treated ingot had coarser graphite (Fig. 102), and both ingots showed flotation of kish graphite to the tops of the ingots. The general graphite structure became coarser in the vicinity of kish graphite in both samples.

Hypo-eutectic melts using the purer material, carried out in the same manner but air-cooled, gave mixed structures of very fine undercooled graphite (Fig. 103) and very small spherulites distributed in an interdendritic pattern (Fig. 104). At high magnifications the spherulitic structure could just be discerned, and the spherulites were usually seen to be associated with very fine flakes (Fig. 105). The addition of calcium silicide to melts of this pure material

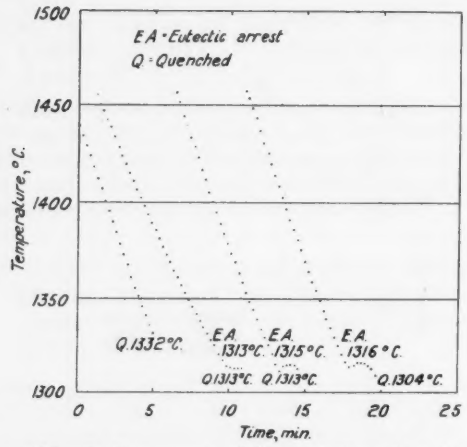


FIG. 136—Cooling curves of four cobalt–carbon ingots melted in graphite crucibles and quenched at different temperatures during solidification. (For structures *see* Figs. 106 to 110)

followed by air-cooling gave mixed structures of flake and undercooled graphite, as has been illustrated in Fig. 99.

Four cooling-curve experiments were carried out on 6-g. ingots, using the powdered cobalt melted in a graphite crucible. The first ingot was quenched in water before the eutectic arrest was reached, the second about half-way through the eutectic arrest, the third at 2° C. below the maximum temperature during the arrest, and the fourth at 12° C. below the maximum temperature of the arrest. The structures of these ingots are illustrated in Figs. 106, 107, 108, and 109, respectively, and the corresponding cooling curves are given in Fig. 136. These experiments do not reveal anything positive about graphite formation in cobalt–carbon alloys. The results are partly marred by the undoubted inoculating influence exerted by primary kish graphite. It is also impossible to say whether solidification was complete in the sample quenched at 2° C. below the maximum arrest temperature (Fig. 108). However, in Fig. 109, which shows the sample quenched at 12° C. below the maximum arrest temperature, an interesting feature is seen, a number of dark areas of very fine graphite being visible. It is doubtful whether this formed during solidification, and it does suggest a process similar to that in iron–carbon–silicon and nickel–carbon alloys. At high magnifications small amounts of cobalt carbide were found associated with this fine graphite (Fig. 110). In view of the fact that this alloy solidified with a largely coarse graphite structure it is conceivable that the alloy was not completely solid at the moment of quenching, although the cooling curve would seem to indicate otherwise.

Occurrence of Cobalt Carbide

The above experiments seemed to indicate that cobalt carbide is slightly more stable than nickel carbide. According to Ruff and Keilig[105] this carbide cannot be detected even in quenched samples. Meyer[106, 107] found that cobalt reacted with carbon monoxide to give, at 500–800° C., a carbide, Co_3C. The crystal structure of this carbide was found to be similar to that of Fe_3C. The unit-cell dimensions for these two compounds are compared below :

Axis	Co_3C (Meyer)	Fe_3C (Lipson and Petch)
a	4·52 Å.	4·5144 Å.
b	5·08 Å.	5·0787 Å.
c	6·73 Å.	6·7297 Å.

In view of the great similarity of the iron–carbon, the cobalt–carbon, and the nickel–carbon alloys which has been demonstrated and will be further buttressed, it seems strange that the structure of nickel carbide should be reported as close-packed hexagonal, $a = 2{\cdot}646$ Å. and $c = 1{\cdot}636$ Å., by Jacobson and Westgren[71] who prepared the compound from nickel powder and carbon monoxide.

It was found that by melting 6 g. of pure cobalt powder in a graphite crucible, heating to 1500° C., slowly cooling to 1400° C., and then quenching rapidly in water at about 15° C., cobalt carbide in the massive eutectic form and the Widmanstätten form could be obtained (Fig. 111). The samples always contained some graphite. The Widmanstätten carbide was undoubtedly deposited from the solid solution during the quench. It will be recalled that nickel carbide was obtained only in the Widmanstätten form by similar treatment. It would appear that cobalt carbide has a stability or persistence midway between that of iron carbide and nickel carbide. In other respects also, the properties of cobalt–carbon alloys lie midway between those of the iron–carbon (or rather iron–carbon–silicon) and the nickel–carbon alloys.

Metallography of Graphite in Cobalt–Carbon Alloys

Graphite flakes in the cobalt-carbon alloys exhibited the same optical properties as those in iron–carbon–silicon and nickel–carbon alloys. Between crossed nicols they lit up four times on revolving the stage through 360°, and under plane-polarized light they exhibited strong reflex pleochroism. The complex internal structures found in the graphite flakes of the other alloys were also present in the cobalt–carbon alloys. This is illustrated in Fig. 112.

Effect of Various Additions to Cobalt–Carbon Alloys

Additions of calcium silicide to cobalt–carbon alloys, even of the lowest sulphur content available, did not give nodular structures as with the nickel–carbon alloys but tended to give a coarsening of the structure. Additions of 0·5% of all the alloys listed in Table IV were made to hypereutectic cobalt–carbon melts, using graphite crucibles, and the results are summarized in Table V.

It is perhaps significant that every alloy containing magnesium gave some nodules and that the calcium–magnesium alloy was completely effective. Also noticeable was the fact that both additions containing lead gave coarse graphite. Fig. 113 shows the completely nodular structure of the cobalt–carbon alloy treated with calcium–magnesium; Fig. 114 shows the structure of the central regions of the ingot treated with calcium–copper; Fig. 115 shows the nodules at the top of this ingot (Figs. 114 and 115 are also typical of the ingot treated with strontium–magnesium); Fig. 116 shows the structure of the ingot treated with calcium–lead.

TABLE V—*Effect of Various Alloy Additions on the Structure of Hypereutectic Cobalt–Carbon Alloys*

Alloy Addition	Structure of Cobalt–Carbon Ingot
80% CaSi ...	Fine eutectic undercooled graphite. Flake kish floated to the top
Ca–Cu ...	A few nodules and remainder undercooled. Nodular kish at the top
Ca–Zn ...	Fine undercooled graphite with flake kish floated to top
Ca–Pb ...	Very coarse flake graphite
Sr–Mg ...	A few nodules and remainder undercooled. Nodular kish at top
Ba–Mg ...	Patches of nodules and patches of undercooled
Ba–Pb ...	Very coarse flake graphite
Ca–Mg ...	Completely nodular

A cobalt–magnesium alloy was prepared containing approximately 30% of magnesium, and a hypereutectic melt was treated with 0·5% of this alloy. In the vicinity of the cobalt–magnesium compound in the resulting ingot, nodular graphite was produced. Again it appeared that alloyed magnesium will enable the graphite to form in the spherulitic pattern.

Temperature of Eutectic Arrest in Cobalt–Carbon Alloys

It will be recalled that Boecker[102] found the eutectic arrest at 1302° C. and Hashimoto [104] found it at 1315° C. In the work reported here the value 1321° C. was obtained for the untreated 40-g. ingot and, when using 6-g. ingots, 1313° C., 1315° C., and 1316° C. were the temperatures obtained. In all these cases the purer cobalt powder was used. The low value agrees with that of Hashimoto. The 40-g. ingot gave a coarse flake graphite structure, whereas the 6-g. ingots contained some finer graphite structures. If the finer graphite structures arise from undercooling, it is normal that the eutectic-arrest temperature should be lower than that of alloys with coarse flake graphite. It was suspected, however, that sulphur, apart from its effect on the graphite structure, would influence the temperature of the eutectic arrest.

The authors possessed no information about the cobalt–carbon–sulphur diagram, but the cobalt–sulphur binary diagram is well established.[108] It shows the occurrence of a cobalt/cobalt–sulphide eutectic solidifying at 879° C., with 26·5% of sulphur. It might be expected from this that sulphur additions to cobalt–carbon alloys would have a big effect in lowering the temperature of the eutectic arrest. In order to check this and to obtain a clearer knowledge of the true eutectic temperature in cobalt–carbon alloys, a number of cooling curves were taken on ingots weighing about 40 g., made up from cobalt powder and melted in graphite crucibles. Each ingot was slowly cooled from this temperature for the cooling curve. In each case a new thermocouple consisting of platinum and 10% rhodium–platinum wire of 0·25-mm. dia. was used. The e.m.f. of the couple was measured with a null-deflection potentiometer, and readings were taken to 0·005 millivolt. A summary of the more important of these experiments, with the structures obtained and the treatments used, is given in Table VI, and the corresponding cooling curves are given in Fig. 137.

From these results it is obvious that the sulphur content and the form of the graphite have a big influence on the temperature of the eutectic arrest in cobalt–carbon alloys. When coarse graphite is produced with a low sulphur content by means of the oxidized alkaline-earth alloys, the eutectic arrest occurred at 1319–1321° C., these being the lowest and highest values obtained by this treatment. This is in fair agreement with the previously obtained value of 1321° C. for an untreated ingot giving coarse graphite. When the coarse graphite was produced by the addition of sulphur, the eutectic arrest occurred at a considerably lower temperature, depending upon the sulphur content; *e.g.*, with 0·3% of sulphur it occurred at 1300° C. When an undercooled graphite structure was produced by melting with

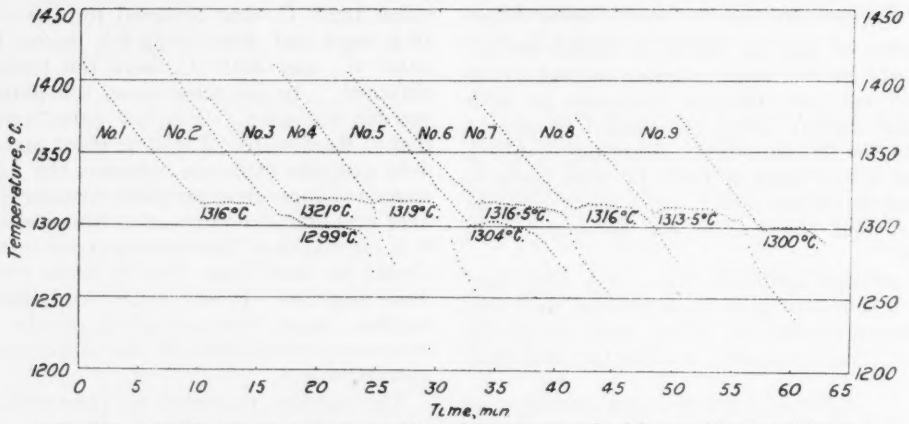


FIG. 137—Cooling curves of 40-g. cobalt–carbon ingots melted in graphite crucibles and treated in various ways to show the influence of sulphur content and graphite structure on the temperature of the eutectic arrest. (*See* Table VI and Figs. 117 to 119)

molybdenum sulphide, which added neither sulphur nor molybdenum to the final ingot, the arrest was at 1304° C. Fig. 117, taken from an ingot treated with strontium–magnesium alloy by ignition, shows a typical coarse graphite structure produced in this series, and Fig. 118 shows the undercooled graphite structure in the ingot treated with molybdenum sulphide.

It is felt that it can be concluded that in cobalt–carbon alloys of the same sulphur content, cooled at the same rate and similar in other respects, a coarse eutectic graphite structure may form at a temperature as much as 15° C. higher than that at which an undercooled graphite structure will solidify. This is in agreement with the idea of undercooling in these alloys and is exactly analogous to the behaviour of cast irons.

The structure produced by the addition of tellurium to a cobalt–carbon alloy is interesting. The authors have no knowledge of the cobalt–tellurium equilibrium diagram, but it may be assumed to be similar to the cobalt–sulphur diagram. If this is so, then tellurium additions might be expected to lower the temperature of

TABLE VI—*Temperature of Eutectic Arrest in Cobalt–Carbon Alloys*

Ingot No.	Treatment	Analysis	Structure	Temperature of Eutectic Arrest, ° C.
1	None	S, 0·012%	Mostly coarse flake with few patches of undercooled	1316
2	0·2% of Te added when molten	...	Abnormal undercooled graphite suggesting carbide breakdown. A few nodules	1299
3	0·5 g. of Sr–Mg alloy ignited on surface of molten alloy, and oxide then stirred in	S, 0·012%	Very coarse graphite	1321
4	0·5 g. of Ca–Mg alloy ignited on surface of molten alloy, and oxide then stirred in	S, 0·013%	Very coarse graphite	1319
5	0·2 g. of powdered molybdenum sulphide mixed with cold charge	S, 0·014%; Mo, nil	Kish + primary metal dendrites + undercooled graphite	1304
6	None	...	Coarse flakes + small patches of undercooled graphite	1316·5
7	0·2% of S, as roll sulphur, included in cold charge	S, 0·166%	Coarse graphite	1316
8	1% of Mn in charge. Treated with 0·3% of S when molten	S, 0·201%	Coarse graphite. Only manganese sulphide present	1313·5
9	0·5% of S, as roll sulphur, in cold charge	S, 0·320%	Coarse graphite. Cobalt sulphide	1300

the cobalt–carbon eutectic arrest. It has been shown that tellurium gives coarse graphite in nickel–carbon alloys, and that in cast irons it introduces a peculiar chilling effect arising from its carbide-stabilizing properties. The tellurium-treated cobalt–carbon alloys showed a structure (Fig. 119) which at low powers resembles undercooled graphite, but at high magnification was seen to be of the " mesh " variety, which has been described by one of the present authors elsewhere.[92] The " mesh " graphite structure is, in any given cluster, a continuous reticulated membrane of graphite. It has been suggested that it is produced in cast irons by the decomposition of a white iron after solidification, owing to a decrease in an external carbide-stabilizing influence (in cast irons this depended upon the

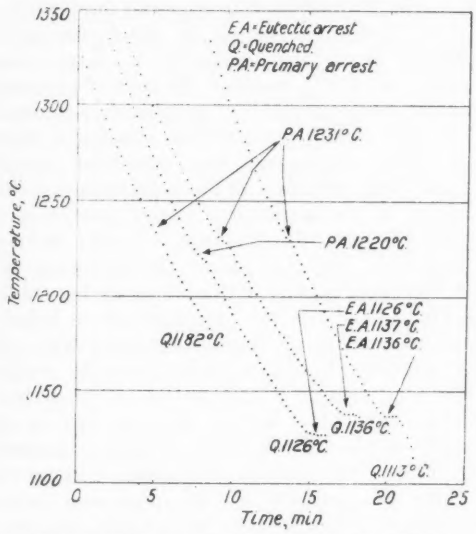


FIG. 138—Cooling curves of 7-g. cast-iron ingots containing 10% of silicon and quenched at various stages during solidification. (*See* Table VII and Figs. 120 to 122)

effect of temperature upon the reaction $MnS + Fe \rightleftharpoons FeS + Mn$). In the case of tellurium-treated cobalt–carbon alloys, it is possible that tellurium behaves as a carbide stabilizer, giving a eutectic carbide which decomposes after solidification. This is indicated to some extent by the simultaneous occurrence of a few spherulites in this structure. Furthermore, it has been found that cobalt–carbon alloys are susceptible to carbide stabilization by the addition of other carbide-stabilizing elements—considerable amounts of cobalt carbide have been obtained by the addition of sulphur or molybdenum to air-cooled cobalt–carbon ingots. If the addition of tellurium to ingot 2 (Table VI) did cause the solidification of a Co–Co_3C eutectic, then this may have also contributed to its low eutectic arrest temperature.

It is relevant here to point out that undercooled graphite structures which have been produced, during this investigation in hypereutectic cobalt–carbon alloys were always associated with primary graphite crystals and primary metal dendrites. In this respect cobalt–carbon alloys closely resemble the cast irons. In nickel–carbon hypereutectic alloys large areas consisting entirely of undercooled graphite with no primary metal dendrites (Figs. 46 and 53) were easily obtained. No structures of this type have been found in cast irons or cobalt–carbon alloys, but, on the other hand, primary dendrites from both sides of the eutectic were occasionally obtained in nickel–carbon alloys (Fig. 45).

Formation of Undercooled Graphite in High-Silicon Cast Irons

It has been mentioned earlier that the presence of silicon or other graphitizing elements does not appear to influence the ease with which undercooled graphite can be produced in cast irons. Since this form of graphite is a product of carbide decomposition, one might expect factors influencing carbide stability to influence the form of its decomposition products. Hurst and Riley[109] have commented upon the occurrence of a carbide phase in high-silicon irons containing 10% and 15% of silicon. This phase appears to possess stability or persistence higher than one might expect from the mere extrapolation of the influence of silicon from lower values. These investigators, by planimetric measurements, indicated that the carbon content of this phase would be of the order of 2·0–3·5%.

To investigate the mode of formation of undercooled graphite in irons of high silicon content, a number of 0·875-in. dia. bars of the following composition were prepared :

T.C.	Si	Mn	S	P	Ti
1·10%	10·0%	0·72%	0·047%	0·028%	0·47%

The titanium was added to this iron in order to guarantee a fine undercooled graphite structure for relatively slow cooling rates. According to the work of Jass and Hanemann,[16] an iron–carbon–silicon alloy with this silicon content should have a eutectic carbon content of approximately 1·6%, the arrest beginning at about 1180° C.

Small ingots weighing about 7 g. each were machined from these bars and melted in the small silica crucibles which have been described and illustrated earlier. One sample was quenched whilst completely molten, one from a temperature

TABLE VII—*Quenching Experiments on High-Silicon Cast Iron*

Quenched	Temperature of Quenching, ° C.	Temperature of Primary Arrest, ° C.	Temperature of Eutectic Arrest, ° C.	Microstructure
From liquid	...	...	...	Carbide + δ ferrite
Between primary and eutectic arrest	1182	1231	...	,, ,, ,,
Before completion of eutectic solidification	1126	1220	1126	,, ,, ,,
On completion of euctetic solidification	1136	1231	1137	,, ,, ,,
Below eutectic temperature	1113	1231	1136	Undercooled-graphite + δ ferrite

after the primary arrest but before eutectic solidification had begun, one during eutectic solidification, one immediately on completion of eutectic solidification, and one at 23° C. below the eutectic arrest. The cooling curves for the last four of these are given in Fig. 138. A summary of the microstructures and critical points in these four melts is given in Table VII.

The structure of the sample quenched immediately after solidification is shown in Fig. 120, and that of the sample quenched 23° C. below the eutectic arrest is shown in Fig. 121. It may be argued that, from the cooling curve, it was impossible to say the alloy quenched at 1° C. below the arrest was completely solid, but without doubt the eutectic crystallization had occurred without the formation of graphite. Since Fig. 121 shows normal undercooled graphite, it follows that these alloys solidify white and subsequently graphitize to give undercooled graphite. The dark spots seen in the micrograph are titanium carbide, standing in relief.

It should be noted that in no case was the eutectic arrest a horizontal—this is to be expected in alloys of this composition—and the arrest temperatures quoted are approximately the mid-temperatures of the arrest. On these cooling curves it was impossible to assign definite values for the beginning and ending of eutectic crystallization, but the values are considerably lower than those to be expected from the work of Jass and Hanemann. This may be owing to the presence of impurities (notably manganese and titanium) or to the German workers' values being determined on alloys of exactly eutectic composition, or to the fact that they did not obtain undercooling.

THE FORM OF THE CARBIDE IN AUSTENITIC CAST IRONS

The proprietary cast iron Nicrosilal is a heat-resisting iron which, when correctly manufactured, should have a structure of undercooled graphite with carbides in a matrix of austenite. The composition of this alloy should fall within the following range :

T.C. %	Si, %	Mn, %	P, %	Ni, %	Cr, %
1·6–2·6	4–5	0·5–1	<0·15	18–20	2–4

In order to obtain the required undercooled graphite structure, which is good for heat-resistance, it is also desirable to have some titanium, say 0·2%, present. In spite of the overwhelming preponderance of graphitizing elements (silicon and nickel) over carbide stabilizers (chromium and manganese), the structure usually contains some carbides. It has been observed that the form of these carbides varies according to whether the graphite structure is normal flake or undercooled. Fig. 122 shows the structure of a good Nicrosilal casting with undercooled graphite, and Fig. 123 shows the structure of an inferior Nicrosilal casting of similar analysis, but with flake graphite. At high magnifications the carbide in the good sample is seen to be of the acicular pattern (Fig. 124), whilst the carbides in the inferior sample are relatively coarse and massive (Fig. 125). In samples containing undercooled graphite the carbide is the dispersed phase. This feature is not so clear in samples containing flake graphite, but certainly in many spots the carbide is the continuous phase of the eutectic.

Obviously in this case we have eutectic carbide occurring in two differing forms, depending upon whether the alloy has undercooled or not. In this context it is interesting to recall that Norbury[30] postulated the existence of a normal and undercooled white iron.

UNDERCOOLING—A RECAPITULATION OF IDEAS IN RELATION TO THE PRESENT PROBLEM

It has now been clearly demonstrated for a wide range of iron–carbon–silicon alloys, nickel–carbon alloys, and cobalt–carbon alloys, that undercooling can occur. The normal flake graphite forms by the solidification of a eutectic or binary complex of solid solution and graphite, whereas the undercooled structure in all three alloys probably results from the solidification of a

binary complex of solid solution and carbide, which decomposes to give a fine form of graphite at some temperature after complete solidification.

The concept of a supersolubility curve for eutectiferous and other binary mixtures has been developed by Miers [110] and his co-workers for certain organic compounds. Gayler [111] has applied this concept to the aluminium–silicon alloys and has determined the position of the supersolubility curve for this system, using rapidly cooled and sodium-modified alloys. This investigator not only located the hypertectic point, but was able to plot the supersolubility curve in the hypo-eutectic and hypereutectic regions. In many respects the aluminium–silicon alloys possess a remarkable resemblance to the iron–carbon–silicon alloys in that solidification of the eutectic at normal temperatures gives silicon in a lamellar form, whereas, by undercooling, a finely dispersed form exactly similar in appearance to undercooled graphite is obtained. Thus the problem of correlating our knowledge of undercooling in the aluminium–silicon system with that of the iron–carbon–silicon, the nickel–carbon, and the cobalt–carbon systems presents itself. Gayler was able to determine the extent to which undercooling could occur for the primary arrests on both sides of the eutectic composition. This is necessary for the complete establishment of a supersolubility curve, but, so far, the present investigators have been unable to do this for the alloys relevant to the subject of this paper. Wide variations are undoubtedly obtained from a given iron for the temperature of the primary austenite arrest, and cooling curves frequently show undercooling and recalescence for this arrest, but its actual temperature is very susceptible to changes in carbon and silicon content, so that the establishment of a supersolubility curve—if any exists—is a matter of real difficulty. It should be possible to prevent undercooling at the primary arrest by the addition of austenite just before solidification begins. In this connection it is of significance to note that, as far as the solidification of the eutectic is concerned, inoculation apparently occurs only with graphite additions or in the presence of kish graphite, and not with austenite.

In her discussion on the solidification of undercooled hypo-eutectic aluminium–silicon alloys, Dr. Gayler has suggested the following sequence : Aluminium separates at a point on the supersolubility curve, and continues to be deposited at constant or rising temperature ; the silicon content of the remaining liquid is thereby increased until the normal solubility curve is reached, down which aluminium is deposited until the silicon-supersolubility curve is reached, and silicon is then deposited until the hypertectic point is reached. This explanation, if modified for cast irons, would account for the arrest and recalescence noted at the beginning of austenite separation in the cooling curves for some cast irons.

Thus far it seems that the system aluminium–silicon and those under consideration in this paper are analogous. It should, however, be noted that whereas undercooled graphite solidifies as a carbide structure, we have so far not found analogous behaviour in aluminium–silicon alloys. In all the considerations in the literature of undercooling according to the betol–salol example in non-metallic and metallic eutectics, the present investigators have never found any reference to the possibility of unstable compounds solidifying at the hypertectic, with subsequent decomposition after solidification. Is the behaviour of iron–carbon–silicon, nickel–carbon, and cobalt–carbon alloys unique in this respect, or can the concept of the formation of metastable compounds be extended to other systems ? This is an absorbing question the solution of which is not an easy matter. The present investigators have had no success in proving the existence of a metastable phase in aluminium–silicon alloys, but it is suggested that the possibility should be rigorously examined.

One further type of undercooling in metallic systems[112] should be mentioned here. When a system is such that an alloy will begin to deposit compound A from the liquid until, at a peritectic transformation, A reacts with the liquid to give compound B and eventually, when this transformation is complete, a eutectic of compound B and another phase, C, is obtained, the following may occur. In the absence of nuclei of B the alloy will undercool beyond the peritectic transformation, with the deposition of A until a metastable eutectic of A and C is obtained, which may subsequently decompose after solidification to give B. There are various modifications of this sequence for different types of systems, but in every case a metastable eutectic is formed which decomposes after solidification. Examples of this are the zinc–antimony system,[113] the cadmium–arsenic system,[114] and the iron–phosphorus system.[115] In the zinc–antimony system a maximum on the liquidus curve corresponds to Zn_3Sb_2. The compound ZnSb is formed by a peritectic reaction between Zn_3Sb_2 and antimony, but it does not make its appearance in the absence of inoculation. A metastable eutectic of Zn_3Sb_2 and antimony may decompose as follows :

$$Zn_3Sb_2 + Sb \longrightarrow 3ZnSb.$$

In the sense of the formation of a metastable eutectic which decomposes after solidification, this type of undercooling is analogous to that in the iron–carbon–silicon alloys, but in the light of our knowledge of the liquidus surface in these ternary alloys the analogy cannot be exploited to explain undercooling in cast irons.

Summary and Conclusions

A fairly comprehensive account has now been given of a large number of problems which either need explanation or correlation in connection with the formation of graphite in grey cast iron. Ultimately it should be possible to do this in relation to a single unified theory covering all the various phenomena. In spite of the large numbers of debatable points dealt with, it is felt that many more exist and need to be incorporated into the whole as soon as possible. Only by considering the graphite-formation problem in all its many aspects are we likely to achieve a satisfactory general explanation and hence complete metallurgical control of the material.

In this paper a deliberate attempt has been made to present the facts as given in the literature, interwoven with observations made in the course of the work of the present investigators. The actual experimental work reported here covers only a fraction of the total field to be investigated. However, it is felt that certain facts have been uncovered which enable a bolder approach to be made to the problem. No work has been reported by the present investigators on the effect of gas content or slag–metal equilibria. This section will be tackled next, and it is hoped that it will form the subject of a further report in the future.

Whilst observations have been made and experimental evidence produced on a number of diverse issues, the main weight of the paper has been directed along the following lines :

(*a*) To study and develop the analogies existing between the process of graphite formation in cast irons on the one hand, and in nickel–carbon and cobalt–carbon alloys on the other.

(*b*) To confirm the findings of Eash that undercooling in cast irons occurs with the solidification of a white-iron structure which subsequently decomposes after solidification to give undercooled graphite, and to extend this idea to high-silicon irons and to nickel–carbon and cobalt–carbon alloys.

(*c*) To examine the possibility of nodular graphite occurring in as-cast cast irons, *i.e.*, without any heat-treatment.

These three lines of investigation are closely related at several points and cannot be separated for consideration, but the following main conclusions can be drawn and the following suggestions made, bearing in mind all three :

(1) The analogies existing between the formation of graphite in cast irons, the nickel–carbon alloys, and the cobalt–carbon alloys, can be developed to a surprising degree of detail.

(2) A study of graphite formation in nickel–carbon and cobalt–carbon alloys can yield useful information about the same process in the iron–carbon–silicon alloys and in cast irons.

(3) Whilst there are bound to be some differences of detail, it is reasonable to suppose that any general theory of graphite formation in cast irons may be equally applicable to nickel–carbon or cobalt–carbon alloys. Application in this manner may in fact prove a useful test for the validity of a theory. There is an obvious limit to this process, and it must be used with caution.

(4) Nickel–carbon and cobalt–carbon alloys yield graphite structures which may be of the normal flake variety or of the undercooled variety.

(5) In both nickel–carbon and cobalt–carbon alloys, the eutectic giving coarse flake graphite solidifies at a higher temperature than that giving undercooled graphite, provided that the sulphur content is the same in each case.

(6) The presence of sulphur (and/or tellurium in the case of nickel–carbon alloys) in nickel–carbon and cobalt–carbon alloys tends to prevent the formation of undercooled structures giving coarse flake graphite. In this respect these alloys show an interesting resemblance to iron–carbon–silicon alloys. This function of sulphur is not understood, but it behaves in this manner in nickel–carbon alloys, whether present as nickel sulphide or as manganese sulphide.

(7) Nickel–carbon and cobalt–carbon alloys are susceptible to inoculation by graphite.

(8) Nickel–carbon and cobalt–carbon alloys are either directly or indirectly susceptible to inoculation by silica.

(9) The addition of calcium (and also probably barium and strontium) or magnesium can cause the formation of spherulitic nodular structures in nickel–carbon and cobalt–carbon alloys, provided that the sulphur content of the alloys is sufficiently low. The calcium and magnesium may be added to the alloys in a wide variety of forms in combination with other elements if necessary. These structures are produced with greater ease in nickel–carbon alloys than in cobalt–carbon alloys.

(10) Rapid cooling of nickel–carbon alloys can also give nodular graphite structures.

(11) Eutectic and hypereutectic graphite can be obtained in the spherulitic form in nickel and cobalt–carbon alloys.

(12) The existence of nickel carbide (probably Ni_3C) has been shown in nickel–carbon alloys, and cobalt carbide (probably Co_3C) in cobalt–carbon alloys.

(13) From this work it would appear that Fe_3C is more persistent (or "stable") than either Co_3C or Ni_3C, and that Ni_3C is less persistent than Co_3C. In many respects cobalt–carbon alloys show properties midway between those of iron–carbon(–silicon) and nickel–carbon alloys. For instance, nodules are produced with ease in nickel–carbon alloys, with a little difficulty in cobalt–carbon alloys, and only with great difficulty in iron–carbon–silicon alloys. Cobalt–carbon alloys are probably susceptible to carbide stabilization by the addition of elements such as tellurium; this is also the case in iron–carbon–silicon alloys, but not in nickel–carbon alloys. That there should be a gradation in properties going through the series Fe, Co, Ni, is understandable. Whether all the observations which have been made merely relate to differences of degree and are controlled by the similar electronic structure of the elements iron, cobalt, and nickel, it is as yet impossible to say. It is conceivable that a more exhaustive study will reveal qualitative differences in behaviour between these three groups of alloys. If this is the case, the differing crystal structure of nickel carbide compared with the cobalt and iron carbides may be a determining feature.

(14) The effect of calcium silicide additions to nickel–iron–carbon alloys in giving spherulitic structures has been shown to persist for iron contents of up to about 50%, beyond which the normal graphite-coarsening influence is obtained.

(15) The work of Eash in showing that undercooled graphite structures in cast irons result from the decomposition of a white-iron structure has been confirmed and shown to be applicable even in the presence of very high silicon contents. That normal flake graphite forms directly from the liquid has been confirmed by the present investigators on many occasions.

(16) The form of the white-iron structure which is responsible for the formation of undercooled graphite is acicular in appearance and differs from normal white-iron eutectics in this respect.

(17) Undercooled graphite in nickel–carbon alloys most probably forms in a similar manner, *viz.*, a eutectic of a solid solution of carbon in nickel and nickel carbide solidifies, and the nickel carbide decomposes after solidification, to give undercooled graphite.

(18) Similarly, in cobalt–carbon alloys undercooling may occur with the solidification of a eutectic of solid solution and cobalt carbide which subsequently decomposes after solidification.

(19) The freezing point of the eutectic in calcium silicide treated nickel–carbon alloys giving spherulitic graphite is substantially the same as that in alloys giving undercooled graphite.

(20) It is suggested that the action of the elements calcium and magnesium in giving nodular structures in nickel–carbon alloys is due to these elements causing some stabilization of the nickel carbide, so that it decomposes at a lower temperature than that required to give undercooled graphite. A similar explanation would explain the similar effects in cobalt–carbon alloys. This explanation agrees with the finding that rapid cooling also gives nodules in nickel–carbon alloys; under these conditions also the nickel carbide would not have time to decompose until lower temperatures had been reached.

(21) A search has shown that spherulitic nodular structures can also occur in cast iron. Broadly speaking, these occur in two different ways, *viz.*:

(*a*) In hypereutectic alloys which have undercooled. This undercooling may be induced either by rapid cooling or by the removal of "nuclei" from the melt, *e.g.*, by vacuum melting.

(*b*) In hypo-eutectic alloys associated with inverse chill. It has been shown that inverse chill may be classified into two types and that one type possesses the acicular white-iron eutectic structure which normally decomposes after solidification to give undercooled graphite. The nodules occur associated with this form, and it has been suggested that they are produced by the decomposition of the acicularly formed cementite at some temperature below that required to give undercooled graphite. If this is correct then the process may be similar to that occurring in nickel–carbon alloys.

(22) No evidence has been obtained in any of the alloys studied that spherulitic graphite structures form directly from liquid. It appears that a solid carbide must decompose to yield a

spherulitic structure, and this confirms experience in the malleablizing process.

(23) Deliberately to produce spherulitic structures in cast irons it is necessary to cause an iron to solidify white and then for the cementite to decompose after solidification. It has been shown that decomposition of carbide after solidification can be produced by suitable use of the reaction $FeS + Mn \rightleftharpoons MnS + Fe$. Except in special cases, this type of carbide decomposition does not yield spherulites, and then only part of the total graphite content is in this form. Without any provision of nuclei the carbide will decompose to give a " mesh-like " graphite pattern.

(24) The production of completely spherulitic structures in cast iron requires an iron which will undercool to give the acicular white-iron structure on solidification and the presence of an alloying element, which will cause it to persist until a temperature probably within the range 800–1050° C., beyond which the carbide would be permitted to decompose.

(25) For an iron of any given composition with respect to the elements carbon, silicon, manganese, sulphur, and phosphorus, it is possible that two forms of white iron exist, *viz.*, one (the normal white-iron structure) which, once it has formed, has an increasing tendency to persist (stability) with falling temperature, and one which tends to decompose after solidification to give undercooled graphite and which has an acicular-distributed form of eutectic cementite. It is possible that the second form corresponds to the maximum degree of undercooling obtainable in an iron.

It is felt that the conclusions and suggestions listed above are all to some extent interrelated and reveal an aspect of the metallurgy of cast irons not hitherto dealt with. In addition to this, many observations have been introduced into this paper which are normally dealt with individually but which must be generally borne in mind in order to explain the process of graphite formation in cast irons. A few of these may be listed as follows :

(i) The relation between the graphite-refining influences of titanium and zirconium and the form of the minor phases necessary to give the effect (TiC or ZrC is necessary).

(ii) The influence of tellurium.

(iii) The graphite-refining effect of bismuth.

(iv) Effect of sulphur in promoting flake graphite formation.

(v) Inoculation with ferrosilicon.

(vi) Effect of hydrogen in giving flake graphite.

(vii) Influence of superheating.

(viii) Effect of slag composition.

(ix) Formation of inverse chill.

If these features alone could be introduced into one unified theory, a considerable advance would be made. However, many other facts also require explanation.

In the course of this work and whilst writing this paper, liberal use has been made of the terms " nuclei," " undercooling," and " spherulitic structures." The present investigators are impressed that these terms touch on spheres of relatively abysmal ignorance. The term " nucleus " is widely used, but its precise connotation is rarely defined and it may refer to small solid particles, at one extreme, or merely to a physical or chemical condition favouring a particular type of crystallization at a given point, at the other extreme. Between these two concepts we have the idea of " residual lattices " in liquids where some of the atoms may be linked together in a similar manner to that of the solid. These aggregates could provide centres for the initiation of crystallization on subsequent cooling, or alternatively could be completely destroyed by superheating. Some visible solid particles can apparently function to nucleate a melt in the sense of preventing undercooling. It would be interesting to know to what extent the crystal structure of such particles influences their ability to initiate crystallization and to prevent undercooling. At times, however, " nucleation " is spoken of when no visible solid particles are in evidence. It is a useful concept to some extent, but knowledge of the nature of the " nuclei " must precede a fuller understanding of the process of solidification in cast irons, and also in other metallic and non-metallic substances and mixtures.

As far as undercooling during solidification is concerned, published work consists almost entirely of the classical researches of Tammann and his co-workers in Germany and Miers and his co-workers in this country. For the most part these workers have used non-metallic substances for study, and undercooling in metallic systems has only been approached at all systematically in the case of aluminium–silicon alloys and cast irons. Further work on the fundamentals of undercooling in non-metallic and metallic mixtures is urgently required. The process is of immense practical significance in metallurgy, because the useful properties of an alloy may be considerably modified by the promotion or prevention of undercooling during solidification.

The existence of graphite in a spherulitic pattern in iron–carbon–silicon, nickel–carbon, and cobalt–carbon alloys is a unique feature, not

only for graphite but also for metallic systems, as far as the present authors are aware. What determines whether crystallization or the formation of a new phase shall give a spherulitic pattern ? Can spherulites form directly from a liquid or is a constricting condition necessary, such as that existing in a transformation in the solid or the solidification of liquid crystals ? Is it necessary that a particular type of nucleus be present to initiate spherulitic crystallization and, if so, does the crystal structure of this nucleus play a part ? Spherulitic graphite formation during the malleablizing process has been shown by one of the authors to occur only in the presence of iron sulphide, and frequently iron sulphide occurs at the centres of graphite spherulites. Manganese sulphide does not have this influence. Is this owing to a difference in crystal structure ? From the authors' appreciation of the literature on this subject it appears that very little is known, but the subject is of vast importance to the metallurgy of cast iron because it involves the difference between a brittle non-ductile material and one with shock resistance and ductility. Again, one might ask what might be the properties of other alloys if other phases in other systems could be made to crystallize spherulitically. For instance, it is conceivable that aluminium–silicon alloys could be produced with vastly different properties if the eutectic silicon could be obtained in this form. It may be that only substances with a certain type of crystal structure can be made to crystallize spherulitically, and this would limit the possibilities. The field here is rich for further fundamental work.

Finally, the present investigators would like to emphasize the value in physical-metallurgical research of studying a particular type of problem in a number of alloys of different components, rather than in one alloy only. It is felt that this opinion is supported by some of the results presented in this paper. The analogous systems must be chosen judiciously. One element at least must be common to the systems, or one element in each system should possess a common feature, such as electron arrangement, and, better, still, in essentially binary alloys one element should be common in each system and the other elements should possess common features. This has been forecast and demonstrated by the theoretical physical metallurgist for many years.

Acknowledgments

The authors desire to thank the Council of the British Cast Iron Research Association for permission to publish the work recorded in this paper, and the Director, Mr. J. G. Pearce, for his constant interest and encouragement. Work on the formation of graphite in cast iron at the laboratories of the Association was originally begun many years ago by Dr. A. L. Norbury and Mr. E. Morgan, and, although these investigators are no longer members of the staff of the Association, the authors desire to record their indebtedness to them for providing a basis of experience which has been freely drawn upon. Finally, they are indebted to their colleagues for assistance in many aspects, and particularly to Mr. D. Marles of the Association's Metallographical Laboratory.

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HISTORICAL NOTE No. 1

Housing Conditions of Ironworkers in the Sixteenth Century*

By Dr. H. R. Schubert†

THE sixteenth century was a most important period in the history of iron and steel making in Britain, as it was marked by a great expansion which affected almost all branches of the industry. There is abundant information regarding ironworks, output, finance, and the social and economic position of the ironmasters of the time, but much less is known about the living conditions of the workers. A fair amount of documentary evidence offering insight into the wages earned by the labourers in the iron industry exists, but hardly anything is mentioned regarding housing.

When an ironworks was erected at or near an abbey or priory sold into private hands after the suppression of the monasteries, there was no difficulty in finding suitable accommodation for the housing of staff and labourers in the various buildings of such places. When Sir Henry Sidney, the father of the famous poet, had a steelworks erected at Robertsbridge (Sussex) in 1565, the steelmakers were lodged in houses belonging to the former abbey, and this arrangement was still in force in 1609.[1] On the other hand, we sometimes learn of ironworks which had no accommodation for the workers at all. This was the case at the forge near Goodrich (Herefordshire) in 1544, where the ironmaster had to board his four ironworkers in the neighbouring village of Whitchurch, but when he made no

* Received 18th February, 1947.
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[1] E. Straker, "Wealden Iron" (second edition), p. 314. London 1931: G. Bell and Sons, Ltd.

profit from the forge he refused to pay for the living expenses incurred.[2]

However, both the cases mentioned seem to be exceptional; as a rule small houses were built for the workers. When in 1553 George Harper and Thomas Culpeper took a lease of the ironworks in the Forest or park of Southfrith (between Tonbridge and Tunbridge Wells, in Kent) the group of buildings consisted of a mansion called "The Great Lodge" (where the manager of the works—he was also both overseer and accountant—lived), two barns, which were probably used as storehouses for iron and charcoal, one furnace, and one finery. In addition there were seven small cottages (*parva cotagia*), built for the ironworkers and colliers.[3] Despite the somewhat pretentious name of the mansion, there must not be any illusion as to the size of such a building. The sixteenth-century manor-houses often were so small that, when used as farm-houses in later times, they had to be enlarged in order to make comfortable homes for the farmers. In view of this fact, how small the labourers' "small cottages" must have been!

Some additional insight into housing conditions is provided by an inventory made of the manor and ironworks at Ewood, in the parish of Newdigate (Surrey) in 1574; the works were then in the possession of Christopher Darrell, a merchant tailor of London, and were worked by the ironmaster Robert Reynolds, of East Grinstead, in Sussex. According to the inventory there were a furnace, a forge, a coalhouse, and "some cotages and howses with garden plot and yards necessary and mete for the workmen to inhabit in."[4] The differentiation made between cottages and houses may infer that some of the dwellings were larger than the others. More important is that they had yards and gardens, which afforded the occupants some supply of vegetables and perhaps the chance of keeping poultry. There were however, no stables or land for the keeping of any of the bigger domestic animals. Whenever a cow was kept for the labourers as an extra consideration this was done at the expense of the proprietor of the works, as was the case in 1574 at a bloomsmithy belonging to the Earl of Shrewsbury in the Sheffield area. According to the "Memorandum Book"[5] of William Dickinson (the Earl's bailiff of Hallamshire, in Yorkshire), the "blower," the smith, and the horse-driver each had a cow kept "both winter and summer."

It is very difficult to make any statement as to the size of such labourers' cottages and the materials of which they were built, but it is possible to draw some conclusions from documents such as accounts of expenses for building. In the accounts of the ironworks in the Forest of Worth, in Sussex, are charges for erecting a forge-hammer in 1547.[6] One item is: "Making of 3 lytel howses for 2 fyners and 1 founder to dwell in 60 shillings," *i.e.*, £1 for each house, which sum was not more than two-and-a-half times the weekly wage of a Sussex founder of the time, and was three times the wages of a hammerman or a finer. As a comparison, it is interesting to note that the erection of a new storehouse for charcoal at Panningridge, also in Sussex, cost £9 8*s.* 5½*d.* in 1546.[7] The discrepancy in the costs is still more striking when the expense of buying a house for the overseers of the ironworks is considered. When in 1569 the wireworks at Tintern Abbey, in Monmouthshire, owned by the Mineral and Battery Works—one of the earliest joint-stock companies in England—experienced a financial crisis almost approaching bankruptcy, a Company's meeting at Founders Hall in London decided to send two members as overseers to the works and to buy a house for them near the Abbey. According to the Company's account book (now preserved in the Manuscript Department of the British Museum) the acquisition of the house cost £20 10*s.*; in view of the state of its finances, surely the Company was neither willing nor able to pay more than the minimum expenses.

A study of the figures given justifies the conclusion that the labourers' houses mentioned above must have been but very tiny cottages. Such a conclusion is supported by official enquiries into the state of the ironworks erected in Cannock Chase (Staffordshire), in 1610.[8] The investigations disclosed that when Lord Paget had ironworks erected in Cannock Wood in 1560, some of the cottages built for the ironworkers were roofed with clods of earth and with turf. These cottages were considered as being of no further use when the workers should leave them. This completes the picture of the primitive state of labourers' cottages that we are able to gather from the mere accidental remarks in documents of the age.

[2] City Library of Hereford, Court Rolls of the Manor of Goodrich.

[3] Public Record Office, Exchequer Special Commissions, No. 1093; *see also* L. F. Salzman, "British Industries," (second edition), p. 40. London 1923.

[4] E. Straker, "Wealden Iron," p. 453.

[5] Preserved in the Sheffield Central Library.

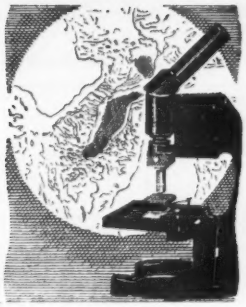
[6] *Archaeological Journal*, 1912, vol. 69, p. 303.

[7] Straker, "Wealden Iron," p. 197.

[8] Public Record Office, Exchequer Special Commissions, No. 4533.

B. I. S. R. A.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION



Report of the Methods of Analysis Committee

The Determination of Phosphorus in Austenitic Chromium-Nickel Steels*

SYNOPSIS

Following detailed investigations, a method has been formulated for the determination of phosphorus in highly alloyed steels containing titanium, vanadium, columbium, zirconium, tungsten, arsenic, and tin. The method is based on a perchloric acid decomposition, which in the early part of the work had been found suitable for carbon and alloy steels not containing the foregoing elements. To this method modifications have been incorporated to overcome the possible interference of these elements, the effects of which have been studied. Where possible, the selected modifications have been those having the widest application, so that undue complication of the basic method has been avoided. The important features of the recommended method are :

(i) *A high nitric acid concentration to prevent the inhibiting effects of titanium and vanadium, and to prevent formation of insoluble zirconium phosphate.*

(ii) *The recovery of occluded phosphorus in the presence of tungsten, columbium, and zirconium by precipitation of magnesium ammonium phosphate from ammoniacal solution, assisted by the addition of arsenate.*

(iii) *Elimination of arsenic and tin by treatment with hydrobromic acid.*

INTRODUCTION

FOLLOWING the publication by the Standard Methods of Analysis Sub-Committee,† of a standard method for the determination of phosphorus in carbon and low-alloy steels, consideration has been given to the problem of phosphorus in highly alloyed materials, particularly austenitic chromium–nickel steels. After preliminary work, which immediately revealed the complexity of this determination, particularly when certain constituents were present, *i.e.*, columbium, zirconium, etc., it was decided that this investigation could best be dealt with by a small Panel empowered to carry out a more detailed examination of the problem. At a Sub-Committee meeting, held on 18th April, 1944, such a Panel, subsequently known as the Phosphorus Panel, was established, its constitution being as follows :

MR. B. BAGSHAWE
The Brown-Firth Research Laboratories
MR. N. D. RIDSDALE
Messrs. Ridsdale and Co., Ltd.
DR. T. B. SMITH
Sheffield University
MR. G. E. SPEIGHT (Convenor)
United Steel Companies, Ltd.

* Paper MG/D/8/48 submitted by the Methods of Analysis Committee of the Metallurgy Division of the British Iron and Steel Research Association, received 29th October, 1946.

This work was carried out by the Phosphorus Panel of the Standard Methods of Analysis Sub-Committee of the Ingots Committee of the Iron and Steel Institute. The duties of this Sub-Committee have now been assumed by the Methods of Analysis Committee of the British Iron and Steel Research Association.

† First Report of the Standard Methods of Analysis Sub-Committee, *Journal of the Iron and Steel Institute*, 1942, No. I, pp. 279–293P.

TABLE I—*Examination of the Effects of Solvent Mixtures on Straight Chromium–Nickel Steels*

Co-operator	Method	Phosphorus Content, %					Phosphorus in Blank, %	Comments
		Steel 4940		Steel 4941		Steel *ASC* 116		
		(i)	(ii)	(i)	(ii)			
B	(2), Gravimetric ...	0·010	...	0·030	...	0·043	Nil	No blanks measurable by direct tests as per method for steel.
	(2A), ,, ...	0·011	...	0·031	...	0·044	,,	
	(2B), ,, ...	0·010	...	0·031	...	0·042	,,	
	(3B), ,, ...	0·012	...	0·030 0·033	0·032	0·043	,,	
R	(3B), Volumetric ...	...	...	...	...	...	...	Abandoned owing to difficultly soluble residues.
	(2C), ,, ...	0·013	0·012	0·030	0·032	...	0·021	High blank (due to HBr) renders these results o less accuracy.
	(2D), ,, ...	0·015	...	0·038	0·039	...	0·006	
S	(1), Gravimetric ...	0·010	...	0·029	...	0·041	$0·000_5$	Slightly low tendency (residue after baking).
	(2), ,, ...	$0·009_5$	...	0·031	...	0·044	$0·001_5$	Decomposition slow. No residue after baking.
	(3A), ,, ...	0·010	...	$0·031_5$	...	$0·043_5$	$0·000_5$	
	(3B), ,, ...	0·010	...	$0·031_5$	...	$0·044_5$	$0·001_5$	Addition of 2 ml. of HNO_3 required after fuming.
								Blanks determined by analysis of reagents.
V	(1), Volumetric ...	$0·011_5$	0·013	0·030	0·030	...	Nil	No blanks measurable by direct analysis of reagents.
	(2), ,, ...	0·010	0·010	0·033	0·032	...	,,	
	(3A), ,, ...	0·010	0·012	0·031	0·031	...	,,	
	(3B), ,, ...	0·010	0·011	0·030	0·033	...	,,	
W	(1A), Gravimetric ...	0·009	...	0·031	...	...	Nil	No blanks measurable by direct analysis of reagents.
	(1A), Volumetric ...	$0·009_5$	...	0·031	...	...	,,	
	(3B), Gravimetric ...	$0·009_5$	...	0·030	...	...	,,	
	(3B), Volumetric ...	0·010	...	0·031	...	...	,,	
Average ...	...	$0·010_5$		0·031		...		...
Range for all methods.	...	0·009 – 0·012		0·029 – $0·032_5$		...		Where 2 or 3 figures are reported, the mean only is considered in the summary.

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The Phosphorus Panel has held frequent meetings during the last two years and has carried out considerable experimentation leading to the formulation of a standard method for the determination of phosphorus in austenitic steels. An abridged account of the work is given in this report, which is presented as the Report of the Phosphorus Panel to its parent body.

It was recognized at the outset that the problem must be considered in two sections, one relating to straight chromium–nickel steels, and the other to similar steels containing titanium, vanadium, zirconium, columbium, tungsten, and tin. Initially, therefore, the work was confined to the investigation of methods for the straight chromium–nickel steels. Part I of the present report gives an account of this section of the work which leads to the formulation of a comparatively simple but accurate method. The investigations of the Panel were extended to cover the more highly alloyed materials and an account of the subsequent work is given in Part II. The method finally recommended by the Panel for such alloyed steels is described in detail in Appendix V of this report.

PART I—THE DETERMINATION OF PHOSPHORUS IN STRAIGHT CHROMIUM–NICKEL AUSTENITIC STEELS

Following a review of the previous work of the Sub-Committee, it was decided to adopt as a basis for future work the method outlined in report "Q. 18/43," reproduced as Appendix I. Alternative solvent mixtures and precipitation conditions had previously been suggested, so that the experimental work was ultimately directed towards an examination of the optimum conditions for the following :

(*a*) Initial decomposition of the sample.
(*b*) Removal of arsenic.
(*c*) Adjustment of precipitation conditions.
(*d*) Composition of the molybdate precipitant.
(*e*) Determination of blanks.

Composition of Solvent Mixture

Complete decomposition of the sample and conversion of phosphorus to orthophosphate are absolutely essential for the successful application of the phosphorus determination. During the

TABLE II—*Examination of the Effects of Solvent Mixtures on Straight Chromium–Nickel Steels*

Co-operator	Method	Phosphorus Content, %						Phosphorus in Blank, %	Comments
		Steel 4940		Steel 4941		Steel *ASC* 116			
		(i)	(ii)	(i)	(ii)	(i)	(ii)		
B	(2), Gravimetric ...	0·011	0·011	0·033	0·032	0·043	0·044	Nil	Slightly low tendencies. Blanks determined by direct analysis of reagents.
	(2D), ,, ...	0·009	0·009	0·030	0·029	0·040	0·041	,,	
	(2E), ,, ...	0·011	0·011	0·030	0·031	0·042	0·042	,,	
R	(2D), Volumetric ...	0·012	...	0·032	0·038*	...	...	0·014	...
	(2E), ,, ...	0·011	...	0·035	0·035*	...	...	0·012	...
	(2E), ,, ...	...	...	0·032	0·032*	...	...	0·006	...
S	(2), Gravimetric ...	$0·009_5$	...	$0·030_5$	...	$0·043_5$	...	$0·001_5$	Blanks determined by direct analysis of reagents.
	(2D), ,, ...	$0·009_5$	...	$0·030_5$	...	$0·043_5$	...	$0·001_5$	
	(2E), ,, ...	$0·009_5$	...	0·030	...	$0·043_5$	...	$0·001_5$	
V	(2), Gravimetric ...	$0·011_5$	0·009	0·032	0·029	...	...	...	...
	(2D), ,, ...	0·009	$0·009_5$	0·032	0·030	...	...	...	...
	(2E), ,, ...	0·010	0·011	0·030	0·030	...	...	...	...
W	(2), Gravimetric ...	$0·009_5$	...	0·031	...	0·044	...	Nil	Blanks determined by direct analysis of reagents. 4940 precipitates slowly. †
	(2D), ,, ...	0·009	...	0·032	...	$0·043_5$	...	$0·0006_5$	
	(2E), ,, ...	0·012†	...	$0·030_5$	...	$0·042_5$	...	Nil	
	(2E), ,, ...	$0·009_5$	...	0·030	...	...	...	...	
Average ...	...	$0·010_5$		0·031		...		...	...
Range for all results ...	...	0·009–0·012		$0·029_5$–$0·033_5$		...		...	Where 2 or more figures are given, a mean only has been considered in the summary.

* 0·2% of arsenic added. † At B.P. in 50 ml. bulk.

progress of this investigation a number of procedures involving a wide variety of solvents and solvent mixtures have been examined:

(1) Aqua regia method (*loc. cit.*, p. 289P).

(2) Perchloric acid method ("Q. 18/43" modified).

(3A) Mixed aqua regia, perchloric acid, and water solvent, completed by conditions described in (1).

(3B) Mixed aqua regia, perchloric acid, and water solvent, completed by conditions described in (2).

Individual co-operators also examined modifications to these methods:

(1) Aqua regia decomposition, followed by treatment as in method (3B).

(2A) Aqua regia, perchloric acid method, similar to (2).

(2B) As (2A), with precipitation in reduced chromium solution.

(2C) Hydrobromic acid solvent, followed by treatment as in (2).

(2D) Dilute aqua regia followed by hydrobromic acid treatment and conditions as in (2).

(2E) Perchloric acid method, similar to (2) in which the first fuming is omitted.

The foregoing methods are specified more precisely in Appendix II.

The phosphorus contents of a selection of austenitic steels were determined using the foregoing procedures and the results obtained were in excellent agreement. From the results given in Tables I and II, and on the score of ultimate accuracy, discrimination between the various procedures was difficult, so that the final selection of a suitable solvent was made on the basis of general applicability to most types of steel. For straight chromium–nickel steels, the recommended method described in Appendix III is substantially the same as method (2A).

During the course of these investigations several interesting features have been revealed. One feature is the importance of ensuring that the phosphorus is converted completely to the correct condition for precipitation by the nitro-molybdate reagent. Attainment of this condition is affected

TABLE III—*Effect of Period of Fuming with Perchloric Acid on the Phosphorus Content of SG 4776 by the Phospho-Molybdate Method*

Sample	Time of Fuming, min.	Condition of Chromium	Phosphorus Content by Analysis, %
SG 4776 (8% of nickel, 18% of chromium, phosphorus content 0·046%)	0	Non-oxidized	$0·026_5$
	1	,,	$0·033_5$
	2	,,	$0·037_5$
	3	Oxidation commenced	$0·037_5$
	4	Partial oxidation	0·045
	6	Complete oxidation	0·046

TABLE IV—*Effect of Baking on the Phosphorus Content of Austenitic Steels by the Aqua Regia Method*

Sample	Accepted Phosphorus Content, %	Conditions of Test	Phosphorus Content by Analysis, %
SG 4435 (*ASC* 103)	0·029	(*a*) Baked lightly...	0·020$_5$ 0·019$_5$
		(*b*) Baked strongly	0·030$_5$ 0·031
F 5815 (*ASC* 110)	0·022	(*a*) Baked lightly...	0·012$_5$
		(*b*) Baked strongly	0·023$_5$ 0·021$_5$
SG 4776	0·046	(*a*) Baked lightly...	0·032 0·034
		(*b*) Baked strongly	0·045 0·047

materially by the presence of chromium, and by the intensity of baking in non-perchloric acid methods, or by the intensity of fuming in the perchloric acid methods. The presence of chromium facilitates the change, but in all steels it is necessary either to bake strongly at 350° C. or to heat to strong fumes of perchloric acid. Decomposition by aqua regia, followed by fuming with perchloric acid also facilitates the change in the absence of chromium, so that this method of decomposition was finally selected. The Panel cannot emphasize too strongly the importance of these points, neglect of which is fatal to accurate determinations of phosphorus. As an example of this danger, the results quoted in Tables III and IV are typical of what may frequently be obtained.

Removal of Arsenic

In view of the possibility of co-precipitation of arsenic in the precipitation of phosphorus by nitro-molybdate, removal of arsenic was considered essential for a standard method. Volatilization as bromide was selected as offering the simplest means of effecting the removal of arsenic. It was found that in the presence of perchloric acid, complete elimination of arsenic is less readily effected by ammonium bromide than by hydrobromic acid, although ammonium bromide is satisfactory in solutions obtained from hydrochloric–nitric acid mixtures. It is necessary in both cases to make the bromide or hydrobromic acid addition to the diluted solution and thereby ensure a reasonable period of evaporation at a moderate temperature for volatilization of arsenic halide. Typical results showing the elimination of arsenic from synthetic test solutions are given in Table V. An addition equivalent to 10 ml. of hydrobromic acid (sp. gr. 1·46) diluted by 20 ml. of water is adequate for the removal of 0·1% of arsenic, and the evaporation should be conducted in the uncovered beaker.

Conditions of Precipitation

Precipitation of ammonium phospho-molybdate is usually carried out in slightly acid solution containing ammonium salts. By close attention to amounts of the reagents and volumes of the solution, it is possible to discard the usual neutralizing and acidifying operations and to proceed with the precipitation as outlined in Appendix I.

Considerable discussion and experimentation have centred on defining a suitable temperature and volume of the test solution before addition of the nitro-molybdate reagent, particularly for

TABLE V—*The Removal of Arsenic from Synthetic Test Solutions of 2 g. of Sample and 0·1% of Added Arsenic, by Treatment with Ammonium Bromide or Hydrobromic Acid*

Composition of Solution	Hydrobromic Acid Addition	Ammonium Bromide Addition	Residual Arsenic, %
Perchloric acid + hydrochloric acid ...	10 ml. (sp. gr. 1·46) ...	...	0·01–0·02
	20 ml. (1 : 1)	...	Trace
	30 ml. (1 : 2)	...	Nil
	...	5 g. + 30 ml. of water ...	0·02
	...	8 g. + 30 ml. of water ...	0·02
Hydrochloric acid + nitric acid + water	10 ml. (sp. gr. 1·46) ...	...	Nil
	...	5 g.	Nil
Hydrochloric acid + nitric acid ...	10 ml. (sp. gr. 1·46) ...	...	0·02
	30 ml. (1 : 2)	...	Nil
	...	5 g. + 30 ml. of water ...	Nil

steels of low phosphorus content. It was suggested that low recoveries may result from the addition of molybdate reagent to 75 ml. of the solution at 80° C., and that 50 ml. of the solution at 100° C. would ensure more uniform conditions and complete precipitation of phospho-molybdate. The comparative test results shown in Tables VI and VII are not indicative of any substantial increase of phosphorus content on precipitation from the higher temperature. It is important to note, however, that with steels of higher phosphorus content, high results are obtained from the higher temperature of precipitation, which suggests the added possibility of contamination with molybdic oxide. Although this effect had not been experienced by all the co-operators, it was decided to recommend the lower temperature of precipitation for all samples, with the introduction of vigorous agitation and an increased period of standing for solutions of low phosphorus content.

Composition of Molybdate Reagent

Two formulæ for the preparation of the nitro-molybdate reagent have been recommended, as experiments have shown that each gives equally satisfactory results (*see* Table VIII). The two solutions are described in Appendix III, one being that recommended in the earlier standard method (*loc. cit.*), the other, being based on solid ammonium molybdate, is considered more representative of modern trends and to have advantages in regard to stability and ease of preparation.

TABLE VI—*The Effect of Temperature on Precipitation of Low Phosphorus Contents*

Method	Phosphorus, %	
	F. 4903	*BCS* A_2
Perchloric acid method (2)... ...	0·008	$0·005_8$
Solution before precipitation: 75 ml. at 80° C.	$0·008_5$	0·005
Allowed to stand for 1 hr.		
Perchloric acid method (2)... ...	0·010	0·006
Solution before precipitation: 50 ml. at 98° C.	$0·009_5$	$0·006_8$
Allowed to stand for 1 hr.		
Direct nitric acid method	0·009	0·006
Solution before precipitation: 50 ml. at 75° C.	$0·008_5$	$0·005_5$
Allowed to stand for 1 hr.		
Accepted figure	0·009	0·007

TABLE VII—*The Effect of Temperature on the Precipitation of Phosphorus by Ammonium Nitro-Molybdate*

Steel	Accepted Phosphorus Content, %	Phosphorus Content (%) on Analysis (Temperature of Precipitation)	
		80° C.	100° C.
A	0·009	0·0085	0·0095
B	0·006	0·005	0·006
C	0·005	0·004	0·005
D	0·043	0·044	0·049

Blank Values

In view of the presence of appreciable quantities of phosphorus in certain grades of perchloric acid, ammonium nitrate, and hydrobromic acid, the necessity for blank determination is strongly emphasized. With the best grades of reagents and using the quantities specified in the method described in Appendix III, this blank approximates to 0·001–0·002% of phosphorus on 2 g. of sample. The blank may be determined separately on known large amounts of reagents, or by treating a sample of electrolytic iron of known low phosphorus content as described in the method. In all cases, the Panel strongly recommends that when using unknown reagents, a standard sample be carried through the method concurrently with the unknown samples.

PART II—THE DETERMINATION OF PHOSPHORUS IN CHROMIUM–NICKEL AUSTENITIC STEELS CONTAINING TITANIUM, VANADIUM, COLUMBIUM, ZIRCONIUM, TUNGSTEN, AND TIN

Following the formulation of the recommended method for straight austenitic steels (Appendix III), subsequent experimental work was directed towards investigating how this method would be influenced by the presence of one or more of the elements mentioned above, and what methods were available for obviating the possible interference. It was decided that the method described in Appendix III should form the basis of a recommended method for the more complex steels which would doubtless necessitate the introduction of suitable modifications to cover the various compositions. In many cases more than one modification has been found suitable for a particular element, and also the same modification has been found suitable for other elements. The final selection of the recommended modification

TABLE VIII—*Phosphorus Determinations using alternative Nitro-Molybdate Reagent Solutions*

Sample	Phosphorus, %	
	Reagent Prepared from Molybdic Oxide	Reagent Prepared from Ammonium Molybdate
SG 4940	0·012	0·011
	0·012	0·011
SG 4941	$0·031_5$	$0·030_5$
	0·032	0·033
	$0·030_5$	0·031
SG 4776	$0·043_5$	0·044
	0·045	0·044
	0·043	$0·042_5$

has therefore been influenced by the latter factor so that the most complex steels may be analysed with the minimum modification to the basic method.

The method of examination has been to apply the draft method (Appendix III) to synthetic samples composed of a straight austenitic steel of known phosphorus content, to which was made an addition of a solution of the element under examination. Comparative tests have also been carried out on austenitic steels containing these additional elements.

The Effect of Titanium

The synthetic additions of titanium were made as a solution of titanium nitrate prepared from potassium titanium oxalate by precipitating titanium hydroxide and re-dissolving in nitric acid. Tests were carried out on steels of low and high phosphorus contents, *i.e.*, 0·010% and 0·030% of phosphorus, with and without the addition of 1% of titanium.

In general, low results for phosphorus were recorded, although varying degrees of interference due to this element were obtained by the individual co-operators. Two possibilities were suggested to account for the interference :

(*a*) Partial hydrolysis of titanium resulted in marked co-precipitation of phosphorus, which was removed from the solution by filtration together with the insoluble silica residue remaining after taking to perchloric acid fumes.

(*b*) Dissolved titanium salts resulted in partial inhibition of the phospho-molybdate precipitation.

The contamination of the silica precipitate was shown to be negligible ; the insoluble residues from nine tests containing 1% of titanium were combined and the resultant phosphorus content determined at 0·004% of phosphorus, thus representing less than 0·0005% of phosphorus for each test. It was therefore concluded that the interference of titanium was one of inhibition of precipitation of the phospho-molybdate. Whilst this effect could be partially reduced by vigorous agitation of the solution during the precipitation, complete precipitation of phospho-molybdate was obtained by any of the following modifications applied immediately before the addition of nitro-molybdate reagents :

(*a*) Addition of 2 ml. of hydrofluoric acid, followed immediately by 1 g. of boric acid.

(*b*) High perchloric acid concentration, followed by vigorous agitation during precipitation. Ten ml. of additional perchloric acid (sp. gr. 1·54) were found to be adequate.

(*c*) High nitric acid concentration and vigorous agitation. Thirty-five ml. of additional nitric acid (sp. gr. 1·42) were found to be adequate.

A selection of typical results obtained on steels containing titanium up to 1% is shown in Table IX, which illustrates the conclusions stated above. The high nitric acid concentration was finally recommended in view of subsequent work which showed that this modification also obviated vanadium interference. The recommended conditions for complete precipitation of phospho-molybdate from solutions containing titanium, which have been obtained by applying the perchloric acid method to titanium steels, include an addition of 35 ml. of nitric acid (sp. gr. 1·42) in a total volume of solution of 100 ml. at 80° C., followed by the addition of nitro-molybdate and vigorous agitation with a glass rod.

An important point to note is that in the presence of the high nitric acid concentration and hexavalent chromium, serious attack of the paper pad may be encountered during filtration of the phospho-molybdate precipitate. This may lead to high results when the volumetric titration is applied in routine practice. This disadvantage is not apparent when the chromium is in the trivalent condition, so that light fuming after the hydrobromic acid treatment is recommended to avoid reoxidation of chromium. If considerable reoxidation of chromium occurs, the chromate should be reduced with sulphurous acid before the precipitation of phospho-molybdate.

The Effect of Vanadium

The effect of vanadium was examined in a manner similar to that of titanium, using synthetic additions of a solution of ammonium meta-vanadate.

It was found that the effect of vanadium in amounts up to 2% on the basic method (Appendix III) was similar to that of titanium. Whilst no

TABLE IX—*The Effect of Various Modifications to Overcome the Presence of Titanium*

Steel	Accepted Phosphorus Content, %	Equivalent Titanium Addition, %	Phosphorus Content (%) on Analysis by Method " Q. 14/44 " (Appendix III), using the modifications below								
			None	Vigorous Agitation	HF + Boric Acid Addition*		$HClO_4$ Addition, sp. gr. 1·54		HNO_3 Addition, sp. gr. 1·42		
					(i)	(ii)	5 ml.	10 ml.	20 ml.	30 ml.	40 ml.
SG 4940	0·010	1	0·002 0·006	0·007 ...	0·010 0·009$_5$	0·009 0·011	...	...	...	...	...
SG 4941	0·030	1	0·008 0·024	0·028 0·027	0·030 0·029	0·031 ...	0·024* 0·029$_5$	0·030	0·027 0·030 0·029$_5$ 0·028	0·030 0·029 0·030 0·030	0·030$_5$ 0·030 0·031 0·029
ASC 116...	0·043	1	0·028 0·039	...	0·043	0·042	...	...	...	...	...
BCS 209 (0·59% of Ti)...	0·018	0·40	0·008 0·006	0·016 0·014	0·017 0·019	0·018	0·017 0·017 0·016 0·016	0·018 0·017 0·018 0·018	...	...	...
18% Cr.	0·023	1	0·016	...	0·022 0·022	...	...	...	...	...	...

* Solutions shaken only, not vigorously agitated with a glass rod.

loss of phosphorus with the silica residue could be detected, the inhibiting effect on the precipitation of phospho-molybdate was observed together with slight contamination of the yellow phospho-molybdate precipitate, to which a deep orange colour was imparted. Modifications similar to those adopted for titanium were investigated :

(*a*) Increased perchloric acid concentration.
(*b*) Increased perchloric acid and nitric acid concentrations.
(*c*) Increased nitric acid concentration.

The increased perchloric acid addition was not successful in eliminating the effect of vanadium ; the results were low and the precipitates orange-coloured, so that attention was directed mainly to the effect of increased nitric acid concentration.

It was decided to examine this modification in the presence of both titanium and vanadium, using varying amounts of nitric acid up to 40 ml. in a total volume of 100 ml., and precipitating the phospho-molybdate from initial temperatures of 80° and 100° C. The degree of vanadium contamination of the phospho-molybdate was also studied, and this appeared to vary with the nitric acid concentration and the temperature of precipitation. The normal yellow-coloured precipitate was obtained when approximately 40 ml. of additional nitric acid (sp. gr. 1·42) were present, but in other cases when considerably less

TABLE X—*The Effect of various Modifications to Overcome the Presence of Vanadium and Titanium*

Steel	Accepted Phosphorus Content, %	Equivalent Addition %		Phosphorus Content (%) on Analysis by Method " Q. 14/44 " (Appendix III), using the modifications below								
		Vanadium	Titanium	$HClO_4$ Addition, sp. gr. 1·54 (Temp., 80° C)			HNO_3 Addition, sp. gr. 1·42					
							Temp., 80° C.			Temp., 100° C.		
				10 ml.			20 ml.	30 ml.	40 ml.	20 ml.	30 ml.	40 ml.
				(i)	(ii)	(iii)						
SG 4941	0·030	Nil	Nil	...	...	...	0·030	0·030	0·030	...	...	0·032
SG 4941	0·030	2	Nil	...	...	...	0·029 0·029$_5$	0·030	0·031 0·031	0·030 0·030$_5$	0·031	0·031$_5$
SG 4941	0·030	2	1	...	...	...	0·031$_5$ 0·029* 0·029* 0·029$_5$	0·031 0·030 0·030†	0·032 0·030 0·030‡	0·032$_5$ 0·030$_5$ ...	0·033 0·031 0·030$_5$	0·034 0·031$_5$ 0·031
BCS 209 (0·59% of Ti)	0·018	1	...	0·014	0·016	0·015	...	...	...	...	...	...
,,	0·018	2	...	0·011	0·011	0·011	...	...	...	...	...	...

*†‡ Denote samples analysed for their vanadium contents.

* Vanadium content = 0·0023% of phosphorus.
† ,, ,, = 0·0012% of phosphorus.
‡ ,, ,, = 0·0007% of phosphorus.

nitric acid was added the orange to yellow-orange precipitate was obtained. Where contamination had occurred, it was generally more severe in those tests which had been precipitated from an initial solution temperature of 100° C. The extent of the vanadium contamination was measured by determining the vanadium content of the final lead molybdate precipitate. From the lower nitric acid concentrations vanadium was found in the lead molybdate in amounts equivalent to 0·0025% of phosphorus, which was reduced to 0·0007% of phosphorus from the higher nitric acid concentrations.

It would appear that although correct or only slightly low results were obtained using the lower acid concentrations, this arises most probably from a balance of the errors, the increased yield resulting from vanadium contamination being balanced by incomplete precipitation due to the inhibiting effect of vanadium. A selection of typical results illustrating the foregoing considerations is shown in Table X.

The recommended conditions, therefore, for the precipitation of phospho-molybdate in the presence of 2% of vanadium from test solutions obtained by the perchloric acid method are similar to those for titanium. The necessity for vigorous agitation of the solution is again stressed.

The Effect of Columbium (and Tantalum)

From the first experiment on the effect of columbium in the determination of phosphorus, it was clearly apparent that the interference was of a different nature to that caused by titanium and vanadium. The latter interference was caused by a partial inhibition of the phospho-molybdate precipitation by dissolved titanium and vanadium salts; with columbium, the initial hydrolysis product of columbic acid, which is removed with the insoluble silica precipitate, exerted a co-precipitation effect resulting in a loss of phosphorus approximately proportional to the amount of phosphorus present.

The results shown in Table XI illustrate markedly the serious interference of columbium. The columbic acid hydrolysis product was removed with the insoluble silica after perchloric acid fuming, and it will be seen that the loss of phosphorus in the presence of a given amount of columbium is approximately proportional to the phosphorus content of the solution. That the loss of phosphorus was due entirely to this co-precipitation effect was proved by combining the hydrolysis residues from these three steels and determining the total phosphorus therein. Duplicate recoveries of 0·038% and 0·035% of phosphorus are approximately equal to the sum of the errors due to the presence of columbium in the three steels.

The preparation of the synthetic columbium solution is described in Appendix IV. A careful study of this is of importance in illustrating the difficulties attendant on the 'synthetic solution' method of investigating analytical procedures. Various methods were examined with the object of finding the most satisfactory and convenient procedures for the complete recovery of phosphorus from the columbic acid hydrolysis precipitate:

(*a*) Following extraction of the residue with hydrofluoric acid, the extract was returned to the main solution which was then treated with boric acid to neutralize the interfering effect of hydrofluoric acid.

(*b*) After the removal of silica, the residue was fused with 1 g. of sodium carbonate and extracted in 30 ml. of semi-saturated sodium chloride solution containing 2 g. of sodium hydroxide. After settling, the extract was filtered through a tight pad and the residue washed with semi-saturated sodium chloride solution. The clear filtrate was acidified with 10 ml. of nitric acid (sp. gr. 1·42), the equivalent of 0·10 g. of ferric iron solution added, and the iron and phosphorus were precipitated with ammonium hydroxide (1 : 1). The precipitate was dissolved in nitric acid and returned to the main solution.

(*c*) The residue was fused with 2 g. of sodium hydroxide and extracted in 15 ml. of semi-saturated sodium chloride solution. After heating for 5 min. the extract was allowed to stand for 1 hr., then filtered through a tightly packed pad, and the residue washed with semi-saturated sodium chloride solution. To the clear filtrate, acidified with 10 ml. of nitric acid (sp. gr. 1·42), an addition of ferric iron was made, equivalent to approx. 0·10 g. of iron (*i.e.*, 5 ml. of the initial filtrate obtained when the insoluble columbium–phosphorus compound was first removed). After precipitation with ammonium hydroxide and filtration, the precipitate was washed back into the solution from which the 5 ml. fraction had been taken.

TABLE XI—*The Effect of Columbium on the Determination of Phosphorus in Austenitic Chromium–Nickel Steels*

Steel	Accepted Phosphorus Content, %	Equivalent Columbium Addition, %	Phosphorus Recovered (including blanks), %	Error due to Columbium, %
SG 4940	0·010	Nil	0·014	0·006
,,	0·010	1	0·008	
SG 4941	0·030	Nil	0·033	0·013
,,	0·030	1	0·020	
ASC 116	0·043	Nil	0·046	0·016
,,	0·043	1	0·030	

The attempt to avoid fusion treatments by using the procedure (*a*) failed due to the subsequent further hydrolysis of columbic acid and contamination of the phospho-molybdate. The fusion treatments (*b*) and (*c*) gave satisfactory separation of phosphorus from columbium on synthetic test solutions, and there was no evidence of columbium contamination at later stages. Typical results obtained by applying treatments (*b*) and (*c*) to synthetic columbium steels are shown in Table XII.

Further work indicated that whilst the two methods (*b*) and (*c*) were suitable for synthetic and commercial columbium 18/8 steels, they were not satisfactory for more highly alloyed steels containing elements such as tungsten, titanium, etc., in association with the columbium. With such steels the separation of columbic acid from phosphorus in sodium chloride solution was incomplete, resulting in subsequent contamination of the yellow precipitate with columbic acid. Treatment with hydrofluoric acid and boric acid reduced this later interference to some extent, but not completely, and it was therefore agreed that an alternative procedure would have to be developed and the sodium chloride extraction separation discarded.

It was considered that a procedure likely to succeed was one based on the precipitation of phosphorus as magnesium ammonium phosphate, from a solution in which columbium, tungsten, etc., were stabilized as complex tartrates. The following treatment of the insoluble residue was therefore adopted :

(*d*) After removal of silica, the residue was fused with 3 g. of potassium bisulphate and the melt extracted in 25 ml. of tartaric acid solution (15%). The solution was diluted slightly, treated with 5 ml. of 'magnesia' mixture (15% of $MgCl_2.6H_2O$, 15% of NH_4Cl), and neutralized with ammonium hydroxide (sp. gr. 0·880) to the change point of bromo-cresol purple. After cooling, a further 5 ml. of ammonium hydroxide (sp. gr. 0·880) were added, and the solution vigorously agitated to induce precipitation. After standing overnight, the precipitate was filtered on a paper pad, washed with cold dilute ammonium hydroxide (3 : 97) and dissolved with 10 ml. of hydrochloric acid (1 : 1) into the filtrate obtained from the initial columbic acid hydrolysis residue.

Tests carried out using the above procedure showed that effective fixation of columbium, titanium, tungsten, and zirconium could be obtained in the alkaline tartrate medium. The phospho-molybdate precipitates obtained following this treatment were free from contamination, except in the case of zirconium. As the subsequent zirconium contamination is only indirectly connected with the application of the modification (*d*) above, it is therefore discussed in the section dealing with this element where a provision is made for its elimination.

The investigations also revealed a marked tendency for incomplete precipitation of magnesium ammonium phosphate from the alkaline tartrate solution, particularly of low phosphorus contents, *i.e.*, equivalent to 0·002% of phosphorus as might be found in residues from commercial samples. The results of a series of synthetic tests shown in Table XIII indicate this low recovery which appears to be associated

TABLE XII—*The Effect of Modifications to Overcome the Influence of Columbium*

Steel	Accepted Phosphorus Content, %	Equivalent Columbium Addition, %	Phosphorus Content (%) on Analysis by Method "Q. 14/44." (Appendix III), using the modifications below			
			None	(*b*) Sodium Carbonate Fusion		(*c*) Sodium Hydroxide Fusion
				(i)	(ii)	
SG 4940	0·010	Nil	0·010	...	...	$0·009_5$
SG 4940	0·010	1	0·004	...	...	0·011
SG 4941	0·030	Nil	0·029	...	...	0·029
SG 4941	0·030	1	0·016	0·030 0·030 0·030	0·029 0·031 0·030	0·031 0·032 0·030
ASC 116	0·042	Nil	0·042	...	...	0·043
ASC 116	0·043	1	0·026	...	...	0·044
1·5% Cb Steel	...	Nil	0·008	...	...	0·018

with inefficient 'seeding' from tartrate solution, since other tests on pure ammoniacal phosphate solutions yielded complete recoveries of 0·002% of phosphorus in the absence of tartrate.

TABLE XIII—*Recovery of Phosphorus from Alkaline Tartrate Solution by 'Magnesia' Method*

Equivalent Phosphorus Added, %	Equivalent Phosphorus Recovered, %
0·002	0·0007
0·002	0·0004
0·005	0·0043
0·005	0·0014
0·010	0·0102
0·015	0·0141
0·015	0·0156

Following the suggestion that the low recovery of phosphate from tartrate solution was associated with inefficient 'seeding out,' the problem was eventually solved by the addition of arsenic as arsenic acid. An addition equivalent to 0·4% of arsenic resulted in a substantial precipitate of magnesium ammonium arsenate and a simultaneous quantitative co-precipitation of magnesium ammonium phosphate, even when only traces of phosphorus were present. It is assumed, therefore, that the function of the arsenic addition is to saturate the solution with magnesium ammonium arsenate, thereby causing a depression of the phosphate solubility; the primary arsenate formation then acting as a nucleus for the phosphate precipitation. The relatively large arsenate addition accelerates this precipitation which is normally slow, and whilst standing overnight may appear to be a desirable feature in ensuring complete separation, the available evidence indicates that with vigorous agitation quantitative co-precipitation of magnesium ammonium phosphate occurs within 1 hr. A suitable method of agitation is to close the neck of the flask with a rubber bung, and shake vigorously for 5 min. after addition of the 'magnesia' reagent. The results shown in Table XIV illustrate the efficiency of the modification for ensuring complete precipitation of magnesium ammonium phosphate from ammonium tartrate solution.

The addition of arsenate in this modification involves its subsequent removal before the precipitation of phospho–molybdate. This has been effected by delaying the hydrobromic acid and volatilization treatments until the modification dealing with insoluble residue has been applied, and its phosphorus content recovered and added to the main test solution.

The recommended method, *see* Appendix V, was applied to a complex 18/8 steel containing 1·5% of columbium with 3% of tungsten, 1% of titanium, 1% of vanadium, 0·5% of zirconium, 0·2% of arsenic, and 0·5% of tin. Although this steel was designed to examine the response of the recommended method to a steel containing all interfering elements under consideration, it is of interest to include the results at this stage. The results of several dozen determinations indicated a mean value of 0·038% of phosphorus, the maximum deviation being ± 0·002%. A steel made from the same basic materials but with the elements such as titanium, columbium, vanadium, etc., omitted, was also prepared for comparison purposes, and the result obtained on this steel was 0·036% of phosphorus. This gives a good indication of the validity of the method for the complex alloy steel, since the increment of 0·002% is a reasonable one, as arising probably from the ferro-alloys employed.

The final recommended method, which is given in Appendix V, includes, therefore, descriptions of the modification to be applied to all steels involving the presence of columbium, either singly or in association with tungsten, titanium, etc. The co-presence of zirconium with columbium requires a further slight modification, similar to that for titanium and vanadium above, *i.e.*, strong nitric acid precipitation conditions.

TABLE XIV—*The Effect of Arsenic Addition on the Recovery of Phosphorus as Magnesium Ammonium Phosphate from Tartaric Acid Solution*

Equivalent Phosphorus Added, %	Recovered Phosphorus, %	
	No Arsenic Addition	With 0·4% Arsenic Addition
0·002	0·0007 0·0004	0·002 0·0019
0·005	0·0043 0·0014	0·0055 0·005
0·010	0·010	0·010
0·015	0·014 0·0155	0·016

The Effect of Zirconium

In the experiments on synthetic 18/8 samples the zirconium addition was made as a solution of zirconium nitrate.

The results of initial tests with zirconium steels using the basic method (Appendix III) indicated that in the presence of 1% of zirconium low results for phosphorus were obtained, and that as with columbium the error was caused by loss of the element as zirconium phosphate in the insoluble silica precipitate removed by filtration after fuming. The amount of phosphorus so lost may amount to as high as 40% of the total phosphorus when 1% of zirconium is present.

The effects of modifications similar to those applied in the presence of columbium were examined :

(*a*) After removal of silica the residue was fused with 1 g. sodium carbonate and extracted in boiling water. After settling, the solution was filtered through a pad, and the residue washed with dilute sodium carbonate solution. The filtrate was acidified with nitric acid, 0·10 g. of iron added, and the iron and phosphorus precipitated with ammonium hydroxide (1 : 1). In certain tests the phosphorus content of this precipitate was determined separately, but in others the precipitate was washed back into the main filtrate remaining after the removal of the initial insoluble silica residue containing co-precipitated zirconium phosphate. The normal precipitation conditions were used, but certain tests were also completed using a high nitric acid concentration.

The results obtained with the modification were satisfactory when zirconium was present singly. For these cases a similar modification has been incorporated in the recommended method. As will be described later, however, it is necessary to use the strong nitric acid precipitation conditions to prevent precipitation of such zirconium phosphate as may be formed during the evaporation with hydrobromic acid. Other modifications were investigated as follows :

(*b*) The residue was fused with sodium carbonate and extracted in semi-saturated sodium chloride solution containing sodium hydroxide, the separation being completed as described in the section on columbium.

(*c*) After fusion in sodium hydroxide the residue was extracted in semi-saturated sodium chloride solution and the separation completed as described for columbium.

(*d*) The residue was fused with potassium bisulphate and extracted in tartaric acid solution, the separation being completed as described for columbium.

Procedures (*a*) and (*b*) gave satisfactory results with both low and high nitric acid concentrations while modification (*c*) sometimes gave high results when the silica was not removed. Results by modification (*d*) were satisfactory, and as this is also applicable to columbium it was adopted also in those cases when columbium and zirconium are in combination. Table XV illustrates some of the experimental results obtained with modifications (*a*), (*b*), and (*c*).

When applying modification (*d*) where, on account of the arsenate addition it is necessary to delay the hydrobromic acid treatment until the insoluble phosphorus is recovered and added to the main filtrate, it was found that the phospho-molybdate precipitate obtained after the evaporation of this solution was contaminated with zirconium phosphate. The error varied between 0·003% and 0·009% of phosphorus in the steels examined which contained up to 0·05% of phosphorus and 1% of zirconium.

TABLE XV—*The Effect of various Modifications to Overcome the Interference of Zirconium*

Steel	Accepted Phosphorus Content, %	Equivalent Zirconium Addition, %	Phosphorus % by Method "Q.14/44"	Phosphorus Content (%) on Analysis by Method "Q. 14/44" (Appendix III), using the modifications below									
				(*a*) Sodium Carbonate Fusion Water Extraction			(*b*) Sodium Carbonate Fusion NaCl. NaOH Extraction				(*c*) Sodium Hydroxide Fusion	
				Nitric Acid Addition, sp. gr. 1·42			Nitric Acid Addition, sp. gr. 1·42				Nitric Acid Addition, sp. gr. 1·42	
				Nil	20 ml.	40 ml.	Nil		40 ml.		Nil	40 ml.
							(i)	(ii)	(i)	(ii)		
SG 4940	0·010	1	...	...	...	...	0·011	...	...	...	0·013	...
SG 4941	0·030	Nil	0·030	...	...	...	...	...	...	...	...	...
SG 4941	0·030	1	0·019	$0·030_5$	0·030	$0·030_5$	0·032 0·031	0·030 0·031	0·031 0·031 $0·030_5$	0·030 0·031 $0·031_5$	$0·031_5$ 0·031 0·032 0·032 0·032 $0·032_5$	0·032
ASC 116	0·043	Nil	0·042	...	...	...	...	...	...	...	...	...
ASC 116	0·043	1	$0·031_5$	0·042	$0·042_5$	$0·043_5$	0·043	...	...	...	0·046	...

Vigorous agitation tended to prevent the precipitation of zirconium phosphate and consequently resulted in complete precipitation of the phosphorus as phospho-molybdate.

As it had been shown that the strong nitric acid conditions of precipitation were satisfactory in the presence of titanium and vanadium, it was thought that these conditions might tend to prevent precipitation of the zirconium phosphate. Subsequent tests confirmed this suggestion; and it was found that boiling for 2 min. after addition of 35 ml. of nitric acid was effective in preventing the precipitation of zirconium phosphate, and that the phosphorus was completely precipitated as phospho-molybdate. Results given in Table XVI illustrate these considerations.

The recommended method therefore includes such modifications as are necessary when zirconium is present singly or in combination with other elements. When zirconium is present singly, a simple fusion in sodium carbonate followed by extraction in water is sufficient, the filtered extract being added to the main solution. When present in combination with columbium and tungsten, the magnesia modification in the presence of arsenate must be applied. In all cases when zirconium is present, the high nitric acid conditions as described for titanium and vanadium are necessary.

The Effect of Tungsten

The occlusion of phosphorus in tungstic acid residues separated by normal evaporation processes in acid solution has been noted previously by many investigators. The possibility of this effect occurring in a perchloric acid solution is not so well established, so that it was necessary to consider the effect of the presence of tungsten in austenitic steels on the determination of phosphorus.

Tests were carried out on a series of steels of approximately similar phosphorus content and of increasing tungsten content up to 21%. Low results for phosphorus were obtained and subsequent examination of the insoluble tungsten residues revealed the presence of phosphorus in amounts increasing with the tungsten content. The occlusion of phosphorus was small up to 6% of tungsten, but with 21% of tungsten the loss of phosphorus equalled approximately 40% of the phosphorus content. Experiments were therefore made to examine suitable modifications, particularly for such steels as contain columbium, etc.

The following modifications similar to those found suitable for columbium and zirconium were examined:

(*a*) After ignition and removal of silica with nitric–hydrofluoric acids, the residue was fused with 2 g.

TABLE XVI—*The Effect of Boiling for 2 min. in High Nitric Acid Concentration, to Prevent Precipitation of Zirconium Phosphate*

Steel	Accepted Phosphorus Content, %	Equivalent Zirconium Addition, %	Phosphorus Content (%) on Analysis by Method "Q. 14/44" (Appendix III), using the modifications below						
			(*d*) Fusion in Potassium Bisulphate. Extraction in Tartaric Acid						
			No Nitric Acid Addition		No Nitric Acid Addition, Vigorous Agitation			35 ml. Nitric Acid added, Boiled 2 min., Vigorous Agitation	
			(i)	(ii)	(i)	(ii)	(iii)	(i)	(ii)
F 9232	0·021	0·20	$0·014_5$	...	0·021	0·021	...	...	...
F 9232	0·021	1	$0·015_5$	...	0·018	...	...	0·023	...
F 9232 + 0·020% of added phosphorus ...	0·041	0·20	0·036	...	0·037	...	...	...	...
F 9232 + 0·020% of added phosphorus ...	0·041	1	0·033	...	0·036	0·038	...	0·042	0·041
SG 5227 (Complex alloy, 0·5% of Zr)	...	...	...	...	...	...	...	0·039	...
F 8528 (0·15% of Zr) ...	...	...	...	...	0·013 0·015	0·014	$0·012_5$	0·015	0·016
F 8528 + 0·020% of added phosphorus ...	...	...	0·031 0·032	0·031	0·033 0·033	0·032	0·030	$0·034_5$	0·035

of sodium carbonate and extracted in the minimum of hot water. Any insoluble residue was removed by filtration and washed with water, keeping the volume of the filtrate below 25 ml. After acidifying with dilute hydrochloric acid (1 : 1), the solution was boiled to remove carbon dioxide, cooled, and made just ammoniacal. The phosphorus was precipitated as magnesium ammonium phosphate, dissolved in dilute nitric acid, and re-precipitated as ammonium phospho-molybdate, the determination being completed by the gravimetric process.

(*b*) Similar to modification (*b*) for columbium, except for the fact that the phosphorus was determined separately in the residue, and not returned to the main filtrate.

(*c*) Similar to modification (*d*) for columbium, including the addition of arsenic acid to ensure complete recovery of phosphorus.

(*d*) After the removal of silica, the residue was fused with 2 g. of potassium carbonate, and extracted in 25 ml. of tartaric acid solution (15%). After acidifying with dilute hydrochloric acid (1 : 1) and boiling to remove carbon dioxide, the solution was cooled, and the method completed as in (*c*) above.

lead to contamination of the final ammonium phospho-molybdate precipitate. Some difficulty was also experienced when tungsten and columbium were present in association in the extraction with the sodium - hydroxide/sodium - chloride solution, due to incomplete separation of columbium leading to low results and loss of phosphorus with the unseparated columbium at the ferric hydroxide precipitation stage. In addition, occlusion of tungstic oxide was more evident when columbium was present. In view of these disadvantages, modifications (*a*) and (*b*) were not considered to be satisfactory.

Procedures (*c*) and (*d*) gave satisfactory results for the separation of phosphorus from tungsten-columbium residues, and as procedure (*c*) had also been proved satisfactory for columbium and zirconium this was finally adopted.

The recommended method, therefore, includes modifications to cover the presence of tungsten

TABLE XVII—*The Effect of various Modifications to Overcome the Influence of Tungsten*

Steel	Accepted Phosphorus Content, %	Tungsten Content, %	Phosphorus (%) by Method "Q.14/44"	Phosphorus Content (%) on Analysis by Method "Q. 14/44," (Appendix III), using the modifications below				
				(*a*) Sodium Carbonate Fusion. Water Extraction	(*b*) Sodium Carbonate Fusion. NaCl + NaOH Extraction	(*c*) Potassium Bisulphate Fusion	(*d*) Potassium Carbonate Fusion	
							(i)	(ii)
No. 2	0·020	2·00	0·019 0·019	0·020	0·020	0·021	...	...
No. 6	0·021	6·00	0·017 0·017	0·021	0·020	0·020	...	...
No. 10	0·024	10·00	0·017 0·018	0·023	0·024	0·022	0·023	...
No. 15	0·025	15·00	0·016 0·017	0·025	0·024	0·023	...	...
No. 19	0·024	19·00	0·015 0·016	0·025	0·023	0·025	...	...
No. 21	0·028	21·00	0·017 0·018	0·028	$0·026_5$	0·026	0·026	...
SG 5227 (Complex alloy steel containing 3·0% of Tungsten)	...	3·00	0·014 0·012	...	...	0·038	0·035	0·037

A typical selection of the results obtained is given in Table XVII. It is clear that while procedure (*a*) gives satisfactory results for tungsten steels, it would obviously be unsuitable for steels also containing columbium. Procedure (*b*) also gave satisfactory results but it was observed that with steels of higher tungsten content, tungstic acid tended to be occluded by the ferric hydroxide precipitate. This is an undesirable feature, as it would subsequently

similar to those for columbium. It has not been considered necessary to formulate a separate modification dealing with the presence of tungsten alone, as the recommended procedure which deals also with columbium, etc., is to be preferred.

The Effect of Tin

From theoretical considerations it was not anticipated that tin would have any effect on the determination of phosphorus by the basic method

(Appendix III). Tin, like arsenic, should be completely eliminated at the hydrobromic acid fuming stage, and moreover, in view of the high acid concentrations of the solutions, it would appear unlikely that phosphorus losses would occur due to co-precipitation of tin phosphate in the insoluble siliceous residue or at any subsequent stage of the analysis. Experiments on the basic method confirmed that the presence of tin up to 1·5% did not adversely influence the determination.

TABLE XVIII—*The Effect of Tin on Method "Q. 14/44" (Appendix III)*

Steel	Phosphorus Content, %	Equivalent Tin Addition, %	Phosphorus Recovered, %
SG 4941	0·030	Nil	0·030
,,	0·030	0·50	0·031
,,	0·030	1·00	0·0305
,,	0·030	1·50	0·031
ASC 116	0·043	Nil	0·0425
,,	0·043	1·00	0·0425

A sample of stainless steel *SG* 4941, to which was added 1·5% of tin, was treated as in the recommended method up to the addition of hydrobromic acid and evaporation to fumes. After cooling and re-dissolving in 250 ml. of water, the solution was heated to 50° C. and treated with hydrogen sulphide for 20 min. The precipitate was collected on a paper pad, washed with ammonium nitrate solution (2%) containing hydrogen sulphide, dried, and ignited at a low temperature. An examination of the precipitate showed that less than 0·02% of tin was present.

A series of phosphorus determinations was carried out on steels of known phosphorus content both with and without an addition of tin equivalent to 0·5%, 1·0%, and 1·5%, respectively. The results are shown in Table XVIII, indicating the complete absence of interference in the recommended method for phosphorus in austenitic steels.

The Recommended Method

Following the experimental work, briefly outlined in the foregoing, the recommended method was formulated as in Appendix V. This is based on the perchloric acid method for straight chromium–nickel steels to which have been added certain modifications to meet the conditions arising from the presence of elements frequently found in such steels. In the description of the method, which is necessarily detailed, the modifications may appear to introduce complexities, but in practice the method will be found to be readily applicable to a wide variety of alloy steel qualities.

As a final examination, the recommended method was applied to a series of current alloy steel types. The results obtained by the members of the Panel are given in Table XIX, together with the general composition of the steels examined. It will be noted that the range of compositions covers carbon steels, 18/8 steels containing titanium and columbium, easy-machining alloys containing zirconium, and complex heat-resisting steels containing tungsten, molybdenum, cobalt, and columbium, and that the results are in good agreement.

SUMMARY AND CONCLUSIONS

As a consequence of investigations carried out by the Phosphorus Panel of the Analysis Sub-Committee (Ingots Committee), a method has

TABLE XIX—*The Determination of Phosphorus in a Series of Complex Alloy Steels by the Recommended Method (Appendix V)*

Steel	Approximate Composition, %											Accepted Phosphorus Content, %	Phosphorus Content by different Laboratories, %							
													A		B		C		D	E
	C	S	Cr	Ni	Co	W	V	Mo	Cb	Tl	Zr		(i)	(ii)	(i)	(ii)	(i)	(ii)	(i)	(i)
BCS 209	...	...	18	8	...	...	...	...	...	0·6	...	0·018	0·019	0·019	0·017	0·016	0·017	0·017	0·018	0·018
BCS 220	...	...	4	...	...	7	1·3	4	...	...	...	...	0·026	0·026	0·024	0·024	0·025	0·026	...	0·025
RR 430	...	...	14	14	...	2·5	...	0·5	0·15	...	...	...	0·020	0·020	0·019	0·019	0·020	0·019	0·020	0·022
5392	...	...	13	13	10	1·5	...	2·5	3	...	...	...	0·021	0·022	0·023	0·022	0·020 0·020	0·021	0·019	0·021
F 8528	...	0·6	13	...	...	...	...	...	...	...	0·15	...	0·016	0·016	...	...	0·015	0·016	0·015	0·019
Cb Rustless ...	...	...	18	8	...	...	...	...	1	...	...	...	0·027	0·028	0·025	0·026	0·025 0·027	0·027	0·023	0·024
Carbon steel ...	0·5	...	...	...	...	...	...	...	...	...	...	0·025	0·023	0·024	0·023	0·024	0·025	0·024	0·0245	0·025

been formulated for the determination of phosphorus in highly alloyed steels containing titanium, vanadium, columbium, zirconium, tungsten, arsenic, and tin. The method is based on a perchloric acid method previously standardized by the Panel, which was found to be suitable for carbon and alloy steels not containing the foregoing elements. To this method a number of modifications have been incorporated to overcome the interference of these elements, the effects of which have been studied. Where possible, the modifications selected have been those which have had widest application so that undue complications have been avoided. The important features of the recommended method are:

(1) A high nitric acid concentration has been found to eliminate the inhibition in the precipitation of phospho-molybdate due to the presence of titanium and vanadium, and to prevent formation of insoluble zirconium phosphate.

(2) When tungsten, columbium, and zirconium are present, loss of phosphorus occurs in the insoluble siliceous residues, which must be treated to recover the occluded phosphorus. The recommended treatment is a magnesium ammonium phosphate precipitation from ammoniacal tartrate solution, 'seeding out' being assisted by the addition of arsenate.

(3) The effect of arsenic and tin is eliminated by treatment with hydrobromic acid.

Appendix I—The Determination of Phosphorus in Straight Chromium–Nickel Austenitic Steels. Method "Q. 18/43." (November, 1943.)

Dissolve 2 g. of sample in 15 ml. of perchloric acid (sp. gr. 1·54) contained in a 400-ml. squat beaker. When the vigorous reaction subsides, rinse the sides of the beaker and evaporate to fumes (note 1). Fume the solution until the chromium is well oxidized (note 2), or in the case of carbon steels, until the fuming mass just turns solid.

Cool slightly, add 40 ml. of hydrobromic acid (1 : 2), evaporate, and fume for a further 5–10 min. (note 3).

Cool slightly, take up in 50 ml. of water, and filter through a pulp pad, washing with hot water to maintain the bulk less than 100 ml. (note 4). Add 5 g. of ammonium chloride, warm the solution to 70° C. and add 20 ml. of nitro-molybdate solution. Shake well and allow to stand at 50° C., for 20–30 min. Filter, washing as usual, and complete the determination by the volumetric or gravimetric procedures.

Nitro-Molybdate Solution

Pour 725 ml. of 15% ammonium molybdate solution (slightly ammoniacal) into 275 ml. of nitric acid (sp. gr. 1·42), filter before use.

Notes

(1) Certain stainless steels are not readily soluble in perchloric acid alone. In such cases, the addition of a few millilitres of nitric acid (sp. gr. 1·42) is recommended.

(2) This fuming is absolutely essential for complete conversion of phosphorus to orthophosphate. Low results are obtained by fuming for a restricted period.

(3) With steels of normal arsenic content, the hydrobromic acid treatment may be omitted, and an allowance made. Under the conditions described above, 0·030% of arsenic increases the phosphorus result by the equivalent of 0·003% of phosphorus.

(4) In steels containing tungsten, titanium, columbium, or zirconium, it will be necessary to examine this residue for its phosphorus content.

Appendix II—The Determination of Phosphorus in Straight Chromium–Nickel Austenitic Steels. Outline of Methods of Decomposition Examined.

(1) *Aqua Regia Method*

Dissolve cautiously 2 g. of sample in a mixture of 20 ml. of hydrochloric acid (sp. gr. 1·16) add 20 ml. of nitric acid (sp. gr. 1·42). Evaporate and bake at 350° C. for 30 min. Take up in 40 ml. of hydrochloric acid (sp. gr. 1·16). Dilute, filter, treat with 10 ml. of hydrobromic acid (sp. gr. 1·50), and evaporate to 8 ml. Add 10 ml. of nitric acid (sp. gr. 1·42) and again evaporate to 8 ml.

Cool slightly, add 16 ml. of dilute ammonium hydroxide (1 : 1) and 10 ml. of nitric acid (sp. gr. 1·42). Dilute the clear solution to 60 ml. Heat to 75° C. and precipitate with 35 ml. of nitro-molybdate reagent. Allow to stand for 30 min. and complete the determination either gravimetrically or volumetrically.

(2) *Perchloric Acid Method*

Dissolve 2 g. of sample in 20 ml. of perchloric acid (sp. gr. 1·54) and evaporate to strong fumes for at least 10 min. Re-dissolve in 10 ml. of hydrochloric acid (sp. gr. 1·16), add 30 ml. of dilute hydrobromic acid (1 : 3), and again evaporate to fumes.

Cool, re-dissolve in 25 ml. of water, filter, and adjust the filtrate to 75 ml. Add 8 g. of ammonium nitrate, warm to 80° C. and precipitate with 35 ml. of nitro-molybdate reagent. Complete the determination as above.

(3A) *Mixed Aqua Regia, Perchloric Acid, and Water Solvent (J. L. West, Messrs. Hadfields, Ltd.)*

Dissolve 2 g. of sample in the following mixture which has been previously heated to boiling point :

20 ml. of hydrochloric acid (sp. gr. 1·16).
5 ml. of nitric acid (sp. gr. 1·42).
5 ml. of perchloric acid (sp. gr. 1·54).
20 ml. of water.

Evaporate to dryness and bake at 350° C. for 30 min. Take up in 40 ml. of hydrochloric acid (sp. gr. 1·16) and digest for 10 min. Add 10 ml. of hydrobromic acid (sp. gr. 1·46) and evaporate to 10 ml. Add 10 ml. of nitric acid (sp. gr. 1·42) and evaporate to 8 ml.

Complete the determination as in method (1) above.

(3B) *Mixed Aqua Regia, Perchloric Acid, and Water Solvent (J. L. West, Messrs. Hadfields, Ltd.)*

Dissolve 2 g. of sample in the mixture given in method (3A), and continue until the evaporation to 10 ml. with hydrobromic acid is completed. Add 15 ml. of perchloric acid (sp. gr. 1·54). evaporate, and fume for 10 min.

Cool, re-dissolve in 25 ml. of water and complete the determination as in method (2).

(1A) *Modified Aqua Regia Method*

Dissolve cautiously 2 g. of sample in a mixture of 20 ml. of hydrochloric acid (sp. gr. 1·16) and 20 ml. of nitric acid (sp. gr. 1·42), evaporate to dryness, and continue as in method (3B).

(2A) *Aqua Regia, Perchloric Acid Method*

Dissolve 2 g. of sample in 20 ml. of hydrochloric acid (sp. gr. 1·16) and 20 ml. of nitric acid (sp. gr. 1·42), add 15 ml. of perchloric acid (sp. gr. 1·54), and continue as in method (2).

(2B) *Aqua Regia, Perchloric Acid Method (B. Bagshawe, Brown-Firth Research Laboratories)*

Similar to method (2A) above. Reduce the solution after perchloric acid fuming and neutralize with ammonium hydroxide to slight acidity of nitric acid. Precipitate with nitro-molybdate and finish as in method (1).

(2C) *Hydrobromic Acid Method (N. D. Ridsdale, Bureau of Analysed Samples)*

Dissolve 2 g. of sample and 3 g. of sodium nitrate in 40 ml. of hydrobromic acid (sp. gr. 1·38). Evaporate to paste, add 20 ml. of perchloric acid (sp. gr. 1·54), and fume strongly for 10 min. after the chromium is fully oxidized.

Cool, re-dissolve in 25 ml. of water, and continue as in method (2).

(2D) *Dilute Aqua Regia, Hydrobromic Acid Method (N. D. Ridsdale, Bureau of Analysed Samples)*

Dissolve 2 g. of sample in a hot mixture of 25 ml. of hydrochloric acid (sp. gr. 1·16), 3 ml. of nitric acid (sp. gr. 1·42), and 20 ml. of water. When dissolved, add 14 ml. of hydrobromic acid (sp. gr. 1·38) and evaporate to a paste. Add 20 ml. of perchloric acid (sp. gr. 1·54) and finish as in method (2C).

(2E) *Modified Perchloric Acid Method*

Dissolve 2 g. of sample in 20 ml. of perchloric acid (sp. gr. 1·54). When dissolved, treat with 20 ml. of dilute hydrobromic acid (1 : 3), evaporate, and fume for 10 min.

Cool, re-dissolve in 25 ml. of water, and continue as in method (2).

APPENDIX III—Circularized as "Q. 14/44." (October, 1944)

Solutions Required

A.R. quality reagents are recommended.

All solutions must be filtered before use.

Hydrobromic Acid (1 : 2)—Dilute 330 ml. of hydrobromic acid (sp. gr. 1·46) to 1 litre with distilled water.

Molybdate Reagent—The following alternatives are permissible :

(*a*) Make a suspension of 100 g. of molybdenum trioxide (98 to 100%) in 240 ml. of distilled water. Complete solution of the acid by the addition of 170 ml. of ammonium hydroxide (sp. gr. 0·880). Pour the solution, with repeated shaking, into 1250 ml. of nitric acid (sp. gr. 1·20) containing 0·04 g. of microcosmic salt. Stand, and filter immediately before use.

(*b*) To 1 litre of nitric acid (sp. gr. 1·20), in which are dissolved 240 g. of ammonium nitrate and 0·04 g. of microcosmic salt, pour, with repeated shaking, 1 litre of ammonium molybdate solution (15%). Stand, and filter immediately before use.

The use of old solutions of molybdate reagent

showing heavy deposits of yellow precipitate is not advisable.

Dilute Nitric Acid (2 : 98)—Dilute 20 ml. of nitric acid (sp. gr. 1·42) to 1 litre with distilled water.

Dilute Ammonium Hydroxide (1 : 2)—Dilute 330 ml. of ammonium hydroxide (sp. gr. 0·880) to 1 litre with distilled water.

Ammonium Acetate Solution—Dilute 1 part by volume of glacial acetic acid with 2 parts by volume of distilled water, and neutralize by the addition of ammonium hydroxide (sp. gr. 0·880). Just acidify by the addition of acetic acid.

Ammonium Chloride Solution—Dissolve 25 g. of ammonium chloride crystals in 100 ml. of distilled water.

Lead Acetate Solution—Dissolve 4 g. of lead acetate crystals in 100 ml. of distilled water.

Procedure

Transfer 2 g. of sample to a 400-ml. conical beaker, add cautiously a mixture of 15 ml. of nitric acid (sp. gr. 1·42) and 25 ml. of hydrochloric acid (sp. gr. 1·16), and digest until dissolved.. Add 20 ml. of perchloric acid (sp. gr. 1·54) and evaporate to fumes. Continue the strong fuming (with the beaker covered) for 10 min. after the chromium appears to be fully oxidized. Cool slightly, add 10 ml. of hydrochloric acid (sp. gr. 1·16), and warm gently to obtain a clear solution and to reduce the chromic acid.

Add 30 ml. of dilute hydrobromic acid, and again evaporate to fumes (with beaker uncovered) until the chromium begins to be oxidized. Cool slightly, re-dissolve in 25 ml. of water by warming gently, and filter through a paper pad, washing several times with small quantities of hot water, and collecting the filtrate in a 400-ml. conical beaker.

Adjust the volume of the filtrate to 75 ml., add 8 g. of ammonium nitrate and adjust the temperature to 80° C. Add 35 ml. of cold molybdate reagent, shake until the precipitate forms, and allow to stand on the bench for at least 30 min., or in the case of phosphorus contents below 0·02%, until the supernatant liquid is clear (note 1). Complete the determination as described in the First Report of the Sub-Committee (*loc. cit.*), (note 2).

Notes

(1) With samples of low phosphorus content complete precipitation is facilitated by vigorously agitating the solution for 2 min. with a glass rod, followed by standing for 1 hr.

(2) Blanks must be carried out on the reagents used. Certain grades of perchloric acid and hydrobromic acid may contain appreciable quantities of phosphorus. Freedom from serious contamination can be indicated by carrying out parallel determinations on standard steels of known phosphorus content obtained by an alternative method. More precise knowledge of the phosphorus content of the reagents named above may be obtained by analysis of a relatively large quantity of the reagent, to which is added 0·1 g. of a pure ferric iron salt followed by precipitation of iron phosphate–hydroxide, which may be re-dissolved in acid and the phosphorus recovered by precipitation as phospho-molybdate under normal conditions of precipitation.

Appendix IV—Preparation of Synthetic Columbium Solutions

Two different methods of preparing a stable columbium pentoxide solution for addition to test solutions were employed during the course of this work, and it was noted that the two methods of preparation failed to give a uniform response. The methods used were as follows :

(*a*) Ferrocolumbium (60%) of weight 0·0335 g. was dissolved in 1 ml. of nitric acid (sp. gr. 1·42) and five drops of hydrofluoric acid, and evaporated to dryness. The residue was extracted with 0·1 g. of oxalic acid, and 5 ml. of water, and the extract added to the test solution. This addition was approximately equivalent to a 1% columbium content.

(*b*) The purest obtainable Cb_2O_5, of weight 0·4 g., was fused with the minimum of potassium bisulphate, extracted with water and the precipitated columbic acid recovered, re-fused with potassium bisulphate and extracted with saturated (4%) ammonium oxalate solution, and made up to 50 ml. The addition of 4 ml. of this solution was equivalent to a 1% columbium addition.

With solution (*b*) it was always possible to obtain a quantitative yield of columbium as the hydrolysis product from test solutions to which it was added. On the other hand, the use of solution (*a*) led to incomplete removal of columbium under identical test conditions, and in this case residual columbium showed a tendency to secondary hydrolysis at a later stage, and often the phospho-molybdate precipitates were found to be contaminated with columbic acid. This secondary precipitate carried phosphorus pentoxide leading to incomplete precipitation as phospho-molybdate.

It was shown that the hydrolysis of these initially unseparated traces of columbic acid could be prevented by treatment with 2 ml. of hydrofluoric acid before phospho-molybdate

precipitation, followed by 1 g. of boric acid. This gave the clue to the reason for the different response of the two solutions. The use of hydrofluoric acid to inhibit columbium hydrolysis suggested that the failure to achieve a quantitative initial hydrolysis was also associated with the use of hydrofluoric acid in the preparation of the synthetic solution. There was added confirmation of this in the fact that there was never any difficulty of obtaining complete hydrolytic precipitation of columbium under test conditions from columbium bearing steels as distinct from plain steels which had been treated with synthetic columbium additions.

It is probable that the use of hydrofluoric acid in this way results in the formation of complex fluo-columbates of such stability that their decomposition by perchloric acid fuming is uncertain and incomplete.

APPENDIX V—Recommended Method for the Determination of Phosphorus in High-Chromium Steels containing Tungsten, Titanium, Columbium, Zirconium, Vanadium, Tin, and Arsenic. (Agreed by the Phosphorus Panel, 5th April, 1946.)

Solutions Required

A.R. quality reagents are recommended.

All solutions must be filtered before use.

Molybdate Reagent—The following alternatives are permissible :

(*a*) Make a suspension of 100 g. of molybdenum trioxide (98 to 100%) in 240 ml. of distilled water. Complete the solution of the acid by the addition of 170 ml. of ammonium hydroxide (sp. gr. 0·880). Pour the solution, with repeated shaking, into 1250 ml. of nitric acid (sp. gr. 1·20) containing 0·04 g. of microcosmic salt. Stand overnight and filter.

(*b*) To 1 litre of nitric acid (sp. gr. 1·20), in which are dissolved 240 g. of ammonium nitrate and 0·04 g. of microcosmic salt, pour with repeated shaking, 1 litre of ammonium molybdate solution (15%). Stand overnight and filter.

Old solutions of molybdate reagent showing appreciable deposits of yellow precipitate (after the initial filtration) are not fit for use and must be discarded.

Dilute Nitric Acid (2 : 98)—Dilute 20 ml. of nitric acid (sp. gr. 1·42) to 1 litre with distilled water.

Dilute Ammonium Hydroxide (1 : 2)—Dilute 330 ml. of ammonium hydroxide (sp. gr. 0·880) to 1 litre with distilled water.

Dilute Hydrochloric Acid (2 : 98)—Dilute 20 ml. of hydrochloric acid (sp. gr. 1·16) to 1 litre with distilled water.

Tartaric Acid Solution (15%)—Dissolve 150 g. of tartaric acid in water and dilute to 1 litre with distilled water.

'Magnesia' Solution (*Magnesium Ammonium Chloride*, 15%)—Dissolve 15 g. of magnesium chloride crystals and 15 g. of ammonium chloride in water and dilute to 100 ml.

'Arsenic' Solution—Dissolve 0·25 g. of arsenious oxide (As_2O_3) in 10 ml. of nitric acid (sp. gr. 1·42). Evaporate to low bulk, add 10 ml. of nitric acid (sp. gr. 1·42), again evaporate to 5 ml., and dilute to 100 ml. with distilled water.

Dilute Ammonium Hydroxide Wash Solution (3 : 97)—Dilute 30 ml. of ammonium hydroxide (sp. gr. 0·880) to 1 litre with distilled water.

Dilute Hydrochloric Acid (1 : 1)—Dilute 500 ml. of hydrochloric acid (sp. gr. 1·16) to 1 litre with distilled water.

Bromo-Cresol Purple (0·2%)—Dissolve 0·2 g. of reagent in distilled water and dilute to 100 ml.

Ammonium Acetate Solution—Dilute 1 part by volume of glacial acetic acid with 2 parts by volume of distilled water and neutralize by the addition of ammonium hydroxide (sp. gr. 0·880). Just acidify by the addition of acetic acid.

Ammonium Chloride Solution (25%)—Dissolve 250 g. of ammonium chloride crystals in distilled water and dilute to 1 litre.

Lead Acetate Solution (4%)—Dissolve 4 g. of lead acetate crystals in distilled water and dilute to 100 ml.

Procedure

Transfer 2 g. of sample to a 400-ml. conical beaker, add cautiously a mixture of 15 ml. of nitric acid (sp. gr. 1·42) and 25 ml. of hydrochloric acid (sp. gr. 1·16), and digest until dissolved. Add 20 ml. of perchloric acid (sp. gr. 1·54) and evaporate to fumes. Cover the beaker with a cover glass and continue the strong fuming for 10 min. after the chromium appears to be fully oxidized. Cool slightly, add 10 ml. of hydrochloric acid (sp. gr. 1·16) to effect reduction of the hexavalent chromium (note 1), dilute with 25 ml. of hot water, boil for 5 min., and filter through a paper pad into a 400-ml. conical beaker, washing several times with small quantities of hot dilute hydrochloric acid solution (2 : 98). Reserve the filtrate if treatment of the residue is necessary (notes 2, 3, and 4) and proceed with the residue treatment as described in the appropriate note.

To the filtrate (or to the combined filtrate and acid extract from the insoluble residue treatment (notes 2 and 3) add 10 ml. of hydrobromic

acid (sp. gr. 1·46) and evaporate to fumes (with the beaker uncovered), avoiding re-oxidation of the chromium. Cool slightly, re-dissolve in 25 ml. of hot water and dilute the solution to 75 ml. (note 5).

Add 8 g. of ammonium nitrate, adjust the temperature to 80° C. and add 35 ml. of cold nitro-molybdate reagent, shaking the solution until a precipitate forms. Allow to stand on the bench for at least 30 min., or in the case of phosphorus contents below 0·02%, until the supernatant liquid is clear (note 6).

Complete the determination as described in the First Report of the Sub-Committee (*loc. cit.*) (note 7).

Notes

(1) In the presence of tungsten, columbium, and tantalum, separation of the acid oxides of these elements is not complete by the single evaporation with perchloric acid as described in the procedure. To avoid the interference that would be caused by subsequent separation of the remaining small amounts of these oxides, the following procedure must be applied :

After reduction of the hexavalent chromium, re-evaporate just to fumes, avoiding re-oxidation of the chromium, cool slightly, and extract in 25 ml. of hot water. Boil the solution vigorously for 1 min., simmer at incipient boiling for 5 min. and allow the residue to settle before filtering through a paper pad as described in the procedure.

(2) In the presence of columbium, tantalum, zirconium, or tungsten, the insoluble residue must be treated to recover the phosphorus co-precipitated with the residue from the solution as follows :

Transfer the residue and filter pad to a platinum crucible and ignite at a dull red heat. Cool, moisten the residue with a few drops of water, treat with 2 ml. of hydrofluoric acid and 5 drops of nitric acid (sp. gr. 1·42), and evaporate to dryness.

Fuse the residue with 3 g. of potassium bisulphate until a clear melt is obtained, extract in 25 ml. of tartaric acid solution (15%), and transfer to a 200-ml. conical flask. Add 4 ml. of ' arsenic ' solution and 5 ml. of ' magnesia ' solution, and adjust the volume of the solution to 50 ml. Make the solution just alkaline with ammonium hydroxide (sp. gr. 0·880), using 2 drops of bromo-cresol purple as the indicator, and add 5 ml. of ammonium hydroxide (sp. gr. 0·880) in excess. Cool to 15° C., shake the solution vigorously for 5 min., and stand for 1 hr. The vigorous agitation may be effected either by closing the neck of the flask and shaking, or by whipping the solution with a glass rod. Any precipitate adhering to the glass rod must be removed either by a small piece of filter paper which is returned to the precipitation solution, or by washing in the dilute hydrochlorie acid (1 : 1) used to re-dissolve the precipitate after filtering and washing.

Collect the precipitate on a tightly packed paper pad, washing 5 times with small amounts of cold dilute ammonium hydroxide solution (3 : 97). Dissolve the residue from the pad with 2 washes of 10 ml. of dilute hydrochloric acid (1 : 1) into the filtrate from the initial insoluble residue, and complete the washing of the filter pad with 6 small washes of hot water.

Proceed with the hydrobromic acid treatment as outlined in the second paragraph of the procedure.

(3) When zirconium occurs singly, *i.e.*, not in association with columbium, tantalum, or tungsten, a simplified treatment of the insoluble residue may be employed. Proceed as follows :

Treat the residue for removal of silica as in note (2). Fuse with 0·5 g. of sodium carbonate, extract with boiling water, filter, and wash with hot water. Acidify with dilute hydrochloric acid (1 : 1) and return the solution to the filtrate from the initial insoluble residue.

(4) With samples containing up to 2% of titanium and/or vanadium, but with columbium, tantalum, zirconium, and tungsten absent, treatment of the insoluble residue is not necessary.

(5) In the presence of titanium, vanadium, and zirconium, a strong nitric acid solution is necessary (*a*) to inhibit the retarding influence of titanium and vanadium on the precipitation of the phospho-molybdate, and (*b*) to prevent further separation of zirconium phosphate with the phospho-molybdate. Proceed as follows :

To the solution, the volume of which must not exceed 65 ml., add 35 ml. of nitric acid (sp. gr. 1.42) and dilute to 100 ml. with water. Add 8 g. of ammonium nitrate, heat to 100° C. and boil for 2 min. Cool to 80° C., add 35 ml. of cold nitro-molybdate reagent, and complete as described in the procedure. With this strong nitric acid precipitation the vigorous agitation of the solution with a glass rod during precipitation (note 6) is essential.

(6) With samples of low phosphorus content, complete precipitation is facilitated by vigorously agitating the solution for 4 min. with a glass rod followed by standing for 1 hr.

(7) Blanks must be carried out on the reagents used. Certain grades of perchloric acid and hydrobromic acid may contain appreciable quantities of phosphorus. Freedom from serious contamination can be indicated by carrying out parallel determinations on standard steels of known phosphorus content obtained by an alternative method. More precise knowledge of the phosphorus content of the reagents named above may be obtained by analysis of a relatively large quantity of the reagent, to which is added 0·1 g. of a pure ferric iron salt followed by precipitation of iron phosphate–hydroxide which may be re-dissolved in acid and the phosphorus recovered by precipitation as phospho-molybdate under normal conditions of precipitation.

An Experimental Furnace for the Investigation of Open-Hearth-Furnace Combustion Problems*

Part I—DESCRIPTION OF PLANT

By A. H. Leckie, Ph.D.,† J. R. Hall,‡ and C. Cartlidge‡

SYNOPSIS

An experimental furnace specially designed for the investigation of some of the variables affecting open-hearth-furnace fuel efficiency and rate of melting is described. The serious experimental difficulties associated with operation at melting temperatures have been avoided by conducting the experiments at a lower temperature, the heat transferred to the hearth being measured by means of calorimeters. This has permitted both a higher degree of accuracy in measurement and greater freedom of experiment than would have been possible if the small-scale melting of steel had been undertaken. The experiments have been designed and regard paid to appropriate similarity criteria so that the results obtained at lower temperatures can be applied to melting practice.

I—INTRODUCTION

DURING recent years the problem of correct open-hearth-furnace design and operation has received intensified study owing to the increased price and shortage of fuel, although it has always been of special interest to the steel industry. The large number of variables involved in the study of steelmaking, the technical difficulties inseparable from work with large masses of material at temperatures of 1600° C. and above, and the high capital and operating costs of the plant place great difficulties in the way of comprehensive experimental work on production furnaces. Although much valuable work has been done by way of measurement and analysis of the operation of a large number and variety of furnaces, anything in the way of a planned series of experiments in which the variables could be systematically investigated and to which the normal production of the furnace could be subordinated has always been ruled out on account of the cost that would be involved and the doubt whether all the variables affecting a production furnace could be brought under control. Even the recent large-scale joint research of the United Steel Companies, Ltd., and the British Coal Utilisation Research Association,[1] which is probably the most comprehensive research on production furnaces yet undertaken anywhere, suffered from this limitation in that the range of experimental conditions had to be restricted to limits that would not seriously interfere with the production of good steel.

The obvious answers to difficulties such as these are experiments on a small-scale furnace, but here again certain difficulties arise, chiefly with regard to the reduction of the technical problems while preserving the necessary similarity conditions. Small-scale-model work may be carried out either in the laboratory, using bench-size models, or on a "technological" or semi-works scale using larger plant. An example of the first type of work is the classical investigation of Groume-Grjimailo,[2] who studied the flow of fluids through furnaces, using small models with water and paraffin instead of air and gas. Semi-works-scale furnaces have been used for experimental work at Sheffield University (about 40 years ago), and during recent years at a works in the U.S.A.[3] Perhaps the most comprehensive experimental work done during the last ten years is that of Rummel,[4] who used both laboratory and semi-works-scale models in an investigation of the influence of mixing conditions on gas combustion. A model furnace of similar dimensions to that used in the present work was set up at Oberhausen.

The work of Groume-Grjimailo, although it yielded much useful information of the flow of gases in furnaces, did not include experiments with actual combustion and heat transfer. The experimental open-hearth furnaces at Sheffield and Pittsburgh were designed and constructed for the preparation of experimental heats of steel. In order to attain the required melting temperatures on a small scale, these experimental furnaces were built on lines very different from conventional furnaces, and consequently would be of

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† British Iron and Steel Research Association.

‡ Shelton Iron, Steel and Coal Co., Ltd.

little use for experiments on combustion and heat transfer in ordinary practice.

The work of Rummel could perhaps be regarded as a very complete answer to most questions regarding factors affecting the mixing of gas and air in industrial furnaces. Though to some extent conclusions (1) and (2) would appear to be incompatible, Rummel's most important conclusions, as affecting open-hearth furnaces, are :

(1) A large difference in velocity between air and gas streams improves mixing and combustion.

(2) High velocity of both gas and air improves mixing and combustion.

(3) Excess air improves combustion.

These findings agree with opinions based on observations at production plants. At the same time, Rummel's work does not seem to have become as widely known and accepted among furnace designers and operators as it merits. This may be owing to the following reasons :

(*a*) It is not immediately easy to apply Rummel's large mass of data to production furnaces, which are seriously affected by variations in draught, air infiltration, operating conditions, etc.

(*b*) The models used, including the types of burner, were rather more ideal than can be employed in practice.

(*c*) Rummel's work was primarily an investigation of gas mixing and not of heat transfer to a hearth, which is the practical operator's interest. For instance, Rummel finds that excess air improves combustion, but it must also be known how this excess air affects heat transfer to the hearth by increasing the waste-gas losses.

It was felt, therefore, that there was a definite field for further work on a model furnace specifically directed towards fuel, combustion, and heat-transfer research. Some of the problems requiring investigation are :

(i) Port design.
(ii) Effect of variations of draught.
(iii) Effect of air/gas ratio.
(iv) Effect of flame luminosity and gas quality.

Whilst these questions have all been tackled on production furnaces, the large number of uncontrolled variables have tended to swamp the effect of any one factor. Statistical methods have been of assistance in these investigations,[5] but in many cases a sufficient range of a particular quantity is not available from the results of production furnaces.

There was therefore a definite need for experiments with a small-scale furnace directed towards a study of combustion and heat-transfer phenomena. At an early stage it had to be decided whether the experimental furnace should be designed to melt steel or whether results of importance could be obtained from a simpler installation in which the many practical difficulties associated with the handling of molten steel were eliminated. The original scheme was for a small furnace using preheated gas and air tapped off the regenerators of one of the furnaces in a production shop, but further consideration of this scheme showed that the ensuring of accurate measurements involved great difficulties.

II—Considerations in the Use and Choice of Models

Similarity Principles

In all work with small-scale models it is essential to make a careful study of the "similarity principles" involved, as otherwise the results of model experiments are likely to be misleading. The simplest similarity principle is that of geometrical layout, that is, the model should follow the scale of its full-size counterpart. This is simply stating the obvious. Another well-known criterion of similarity is the "dimensionless group," of which the Reynolds number governing the nature of fluid flow is the best-known example. Thus, if turbulent motion is being studied, the Reynolds numbers under the conditions used in the model and in the full-sized practical application should be, if not identical, of the same order. Other dimensionless groups govern heat transfer[6] ; also, in the study of this subject, similarity must be observed as far as possible in temperature conditions, emissivity, and so on.

It is almost invariably found, however, that it is impossible to satisfy all similarity criteria in a model. Thus, in fluid-flow experiments, satisfaction of the Reynolds criterion may demand a distortion of the geometry, thus disobeying the demand for geometrical similarity. For instance, in experiments with model ships' hulls it is not possible to satisfy both the Reynolds and the Froude criteria in the same experiment.

Since it is seldom possible to satisfy all the required similarity principles in any one series of experiments with models, it has to be decided which principles can be neglected so that the resulting departures from full-scale conditions are either unimportant or capable of estimation or calculation. The investigator must design his experiments accordingly. The similarity principles

to be considered in small-scale experiments on open-hearth furnaces are (*a*) geometric similarity, (*b*) aero-dynamic similarity, and (*c*) thermal similarity.

(a) *Geometric Similarity*

Geometric similarity is important both from the point of view of investigating the effect of the normal port configurations on mixing and flame shape, and the effect of the layout of the working chamber on heat transfer to the bath. Geometric similarity in a model furnace offers no difficulties, except that as brick is the most convenient material to use for a model large enough to satisfy other similarity conditions, the surface roughness, owing to both the natural roughness of the brick and the roughness due to joints, cannot be comparably scaled. This is only of importance as possibly affecting the flow conditions of the gas leaving the gas port, and since, as shown later, the gas stream must be fully turbulent in order to fulfil the Reynolds criterion, additional roughness in the model is no disadvantage.

(b) *Aerodynamic Similarity*

In an open-hearth furnace the flame is a so-called " diffusion flame," *i.e.*, the air and gas mix by diffusion after the gas has left the burner, in contrast with flames in which the air and gas are premixed. In these diffusion flames the speed of mixing of gas and air depends largely on turbulence, and this in turn depends on the Reynolds number. It is important, therefore, that the Reynolds number in the model should be of the same order as that in a production furnace.

Aerodynamic similarity demands some deviations from production conditions. The Reynolds number of a gas flowing in a conduit is defined as $v.d/\nu$, where v = linear velocity, d = hydraulic diameter of conduit and ν = kinematic viscosity of the gas stream. Consequently in a small-scale model, where d is a fraction of full scale, satisfaction of the Reynolds criterion demands an increase in the quantity v/ν. This may be done by increasing the velocity (which would mean undue distortion of geometric similarity) or by decreasing the temperature, since kinematic viscosity falls with temperature.*

The above considerations show that satisfaction of the Reynolds criterion may be best attained by (i) making the model on as large a scale as is practicable and (ii) running with the gas and air at a lower temperature than in the full-scale furnace. This conflicts with the desirability, from the point of view of heat transfer and combustion investigations, of working at steelmaking temperatures. From the points of view of experimental technique and the attainment of the required degree of accuracy in the measurements, work at steelmaking temperatures involves great difficulties, and a balance has to be maintained between the inaccuracy in measurement which is bound to occur owing to the technical difficulties of high-temperature work and the departure from similarity necessarily attached to work under easier temperature conditions.

(c) *Thermal Similarity*

In a production furnace heat transfer takes place by convection and radiation ; the proportion of heat transfer by radiation rises steeply with temperature. Consequently, unless the temperature in the model is the same as in a production furnace, the heat-transfer conditions will not be the same. Also, the velocity of combustion increases with temperature, so consideration of the temperature criterion is necessary from this aspect.

It is also necessary to ensure that the gas rates (*i.e.*, heat inputs) experimented with correspond to the rates used in production furnaces. The similarity criterion for this is not immediately obvious, but that adopted for the model is similarity of rate per square foot of hearth area for gases of similar calorific value. This implies that operation of the model at a temperature lower than that obtained in production furnaces corresponds to any lack of preheat in the gas and air.

Recently Thring[7] has suggested that a dimensionless group which could be applied is the quantity :

$$\frac{Q}{A\,\sigma\,E\,(T_1^4 - T_2^4)}$$

where Q is the heat input per hour,
A is the area of the heat-absorbing surface,
σ is Stefan's constant,
T_1 is the absolute temperature of the heat-emitting surface,
T_2 is the absolute temperature of the heat-absorbing surface,
and E represents the "combined emissivity " of the surface, all in self-consistent units.

This is an extension of the concept of equal heat input per unit of hearth area to include operating temperature. By including Stefan's constant the quantity is made independent of dimensions and therefore a true similarity criterion, but the use of the fourth power of the temperature neglects all heat-transfer effects except that of radiation ; this is probably justifiable for melting temperatures. The possible

* The quantity v/ν cannot be increased by increasing the temperature, since dv/dT is less than $d\nu/dT$.

use of this criterion in connection with the experimental furnace is discussed in an Appendix, but, as has been mentioned previously, the attainment of true thermal similarity without working at steelmaking temperatures is not possible.

A fuller study of the extent to which it is possible to satisfy the various similarity conditions is given in the Appendix.

Choice of Type of Furnace

The foregoing considerations show that the decision had to be made whether the small furnace should be designed to operate at steelmaking temperatures and to melt steel, or whether more useful results could be obtained from a simplified installation. Three alternatives would be: (1) A conventional melting furnace on a small scale, (2) a furnace working at steelmaking temperatures but using precise methods, such as a calorimeter, for measuring the heat transferred to the hearth, and (3) a furnace working at lower temperatures, using a calorimeter. The proposal to melt steel was rejected on the following grounds:

(*a*) The small-scale melting of steel would involve practical difficulties which would seriously conflict with the attainment of scientific accuracy. To reach melting temperatures in a small furnace demands very hard driving, with resulting severe wear and tear on the brickwork. In ordinary works' practice, the difficulty of maintaining port shape and general structural conditions is well known, and with a small furnace such critical dimensions as the area and configuration of the gas port would not remain constant.

(*b*) The use of preheated gas and air would be necessary, and the maintenance of a supply of gas and air at a constant and known preheat (this would be absolutely essential) would scarcely be practicable. The heat-flow-meter results of Chesters and Thring[1] show the enormous variation in downward heat flow occurring within short time intervals under production conditions.

(*c*) The accurate measurement of the useful heat transferred to the bath would be a matter of some difficulty were molten steel to be used as a medium.

(*d*) The melting of steel would not permit of a sufficient range of operating conditions being investigated.

(*e*) The handling of the materials would involve so much labour and ancillary plant as to make excessive demands on the resources available for the research.

The use of steelmaking temperatures in conjunction with a calorimeter was also rejected in the first place on account of considerations (*a*) and (*b*). In later work it is hoped to increase the temperatures to steelmaking levels.

It was therefore decided to construct a scale replica of the upper part of an open-hearth furnace and to work at whatever lower temperature would be attained by the use of a non-preheated clean mixture of blast-furnace gas and coke-oven gas of a calorific value comparable with the gases used in production furnaces. The heat transferred to the bath would be determined by running a known quantity of water through tanks embedded in the hearth of the furnace and measuring the rise in temperature of the water.

The advantages of this method of attack are:

(i) The heat input can be determined with precision, as the cold clean gases used can be accurately measured as regards volume, temperature, and calorific value.

(ii) The useful heat transferred to the bath can be determined from an easily measured quantity of water and an equally easily measured rise in temperature.

(iii) The dimensions of the furnace remain substantially unchanged and alterations are easy to carry out.

As has already been shown, the only serious objections to low-temperature work are the departure from the radiation/convection heat-transfer relationships encountered in practice and the relative slowing-down of combustion at lower temperatures. These objections are, however, not so serious as might at first be thought. The prime necessities of open-hearth-furnace operation are (*a*) rapid combustion and (*b*) rapid transfer of the resulting heat from the flame to the charge. Consequently, if the conditions to give most efficient combustion and maximum heat transfer to the charge are found in the experimental furnace at, say, 1000° C., it is safe to assume that these conditions will, in the main, apply equally well at steelmaking temperatures, bearing in mind that overall rates of combustion and heat transfer by radiation will be substantially greater at the higher temperature.

The experimental procedure is therefore designed in accordance with this argument. Under each set of conditions investigated the heat transfer to the bath is measured over a full range of heat input. This gives a series of "characteristic curves" for each set of conditions, comparable with similar curves obtained from production furnaces by statistical methods of calculation.[5]

III—The Experimental Plant and its Operation

The Furnace and Layout

The furnace was built at the works of the Shelton Iron, Steel and Coal Co., Ltd., by the courtesy of the Management of this firm, who also generously provided all the labour and materials necessary for its construction. It is a $\frac{1}{8}$th linear scale replica of the upper part of one of the existing shop furnaces, but fitted with an easily removable block assembly so that the port blocks can be changed as required. The blast-furnace gas and coke-oven gas used as fuel are drawn from the adjacent blast-furnaces and coke-ovens belonging to the Company. Since it was decided to dispense with preheat, no regenerators are provided and reversing is not required, although the ingoing and outgoing ends of the furnace were made alike for the sake of geometrical similarity.

The principal dimensions of the furnace are :

Hearth length (block to block) ...	8 ft. 3 in.
Hearth width	2 ft. $3\frac{1}{2}$ in.
Hearth area	18·9 sq. ft.
Roof height (maximum) ...	2 ft. $1\frac{1}{2}$ in.
Gas-port area	26 sq. in. maximum
Air-port area	170 sq. in. maximum
Inclination of gas port	10°
Inclination of air port	35°
Gas-uptake area	1 sq. ft.
Air-uptake area (total for two) ...	1·54 sq. ft.
Overall length of furnace ...	16 ft. 9 in.
Overall width of furnace ...	3 ft. 10 in.

The port design specified in the above dimensions is that used in the first series of experiments described in Part II. However, as mentioned above, provision is made for changing the port design to very different lines.

The general layout of the furnace and plant is illustrated in Figs. 1 and 2, and a view of the furnace is given in Fig. 3.* Fig. 1 shows a section of the furnace itself which, in the main, is self-explanatory. It will be noted that both gas and air are brought up through conventional uptakes and that the outgoing end is made similar to the ingoing end, although the furnace is non-reversing. This form of construction is adhered to so as to conform with geometric similarity. The hearth is constructed of a layer of sand, above which is a series of four metal tanks through which water flows at the rate of about 250 gal./hr.† The tanks are divided by partitions so as to ensure that the flow is evenly distributed over the surface of the hearth. Nickel-wound resistance thermometers are inserted at the entrance and exit of each tank, so that the total heat transferred to the hearth tanks and the proportion of heat going to each successive quarter of the surface can be determined. The electrical connections to these resistance thermometers are described later.

Firebrick tiles are laid above the tanks to form

* Reproduced by courtesy of the Staffordshire Sentinel Newspapers, Ltd.

† This was later raised to 450 gal./hr., when doing experiments with high heat inputs.

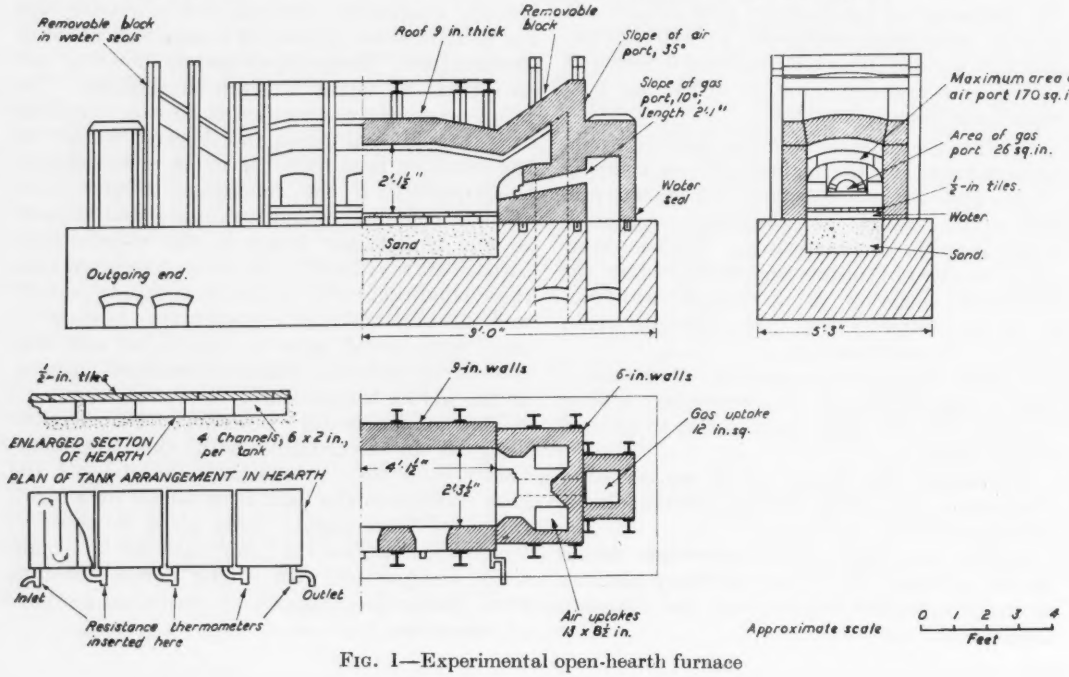

FIG. 1—Experimental open-hearth furnace

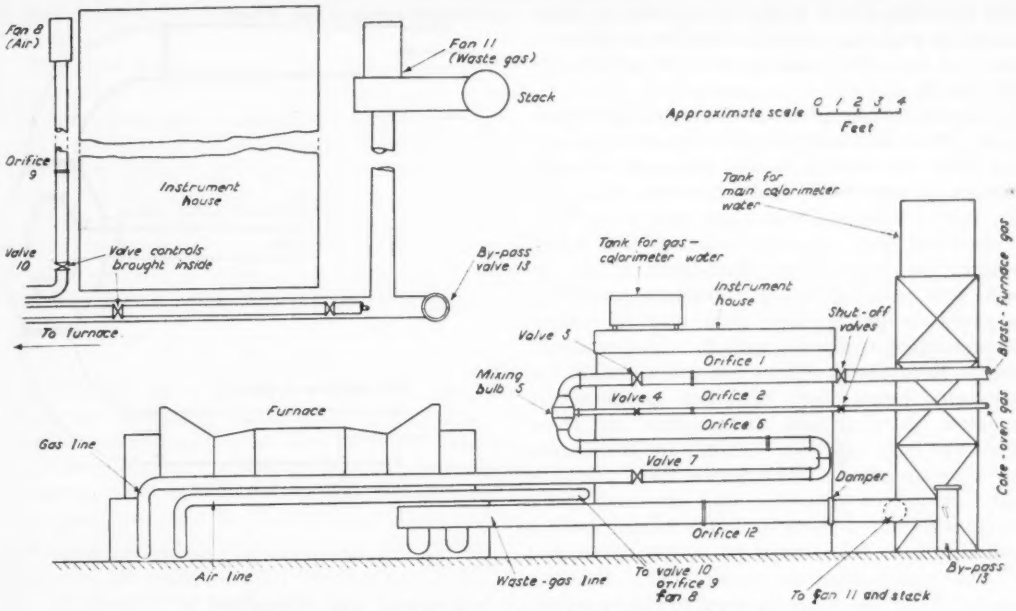

FIG. 2—Experimental open-hearth-furnace site layout

the actual hearth of the furnace. These tiles are for insulating the tank surface; if exposed tanks were used the proportion of heat abstracted into the hearth would far exceed the proportion reaching the hearth in production practice. The thickness ($\frac{1}{2}$ in.) of tiles is calculated so that a proportion of heat roughly corresponding to that in normal practice reaches the hearth, and this is very important. An attempt was made to use tiles made of cast open-hearth slag in order to

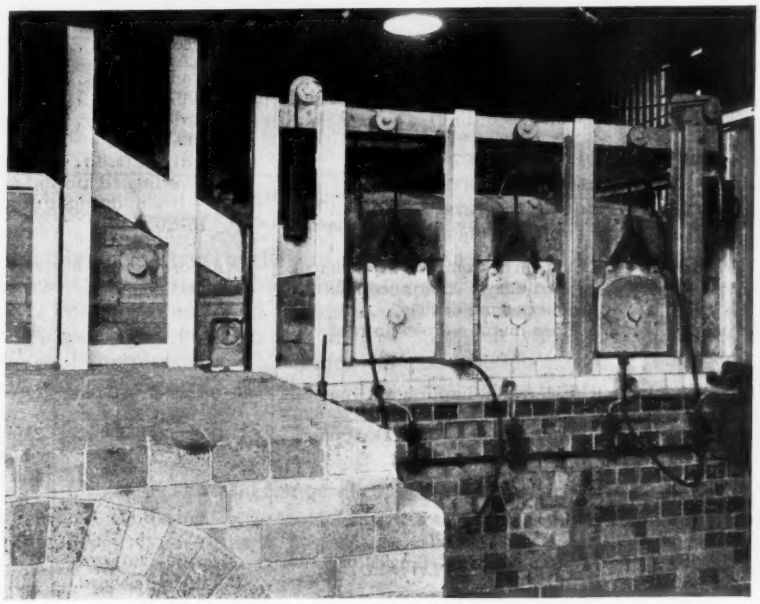

FIG. 3—View of experimental furnace

obtain a surface which might be regarded as more comparable with slag under production conditions. It was not possible, however, to cast sufficiently stable tiles from slag.

The layout of the plant as a whole is shown in Fig. 2. Blast-furnace gas and coke-oven gas are taken from the works supply through pressure governors giving controlled pressures of about 2½ in. W.G. for blast-furnace gas and 4 in. W.G. for coke-oven gas. At a later date it was found necessary to boost the blast-furnace gas to ensure this pressure being uniformly available. The gases are led through main shut-off valves past measuring orifices 1 and 2. The pressure tappings from these orifices are taken inside the instrument house but are not permanently connected to recording instruments, as these orifices are only used for "spot" measurements from time to time. The gases then pass the regulating valves, 3 and 4, and are mixed in a mixing bulb, 5; a section of this mixing bulb is shown in Fig. 4. The mixed gas is then taken past the main gas-measuring orifice, 6. Pressure tappings from this are connected to the recording instrument. After passing through the mixed-gas regulating valve, 7, the gas is led to the furnace.

Air is provided by a fan, 8, passing a measuring orifice, 9, and a control valve, 10. The pressure tappings from the orifice, 9, are connected to a recording instrument. Waste gases are exhausted

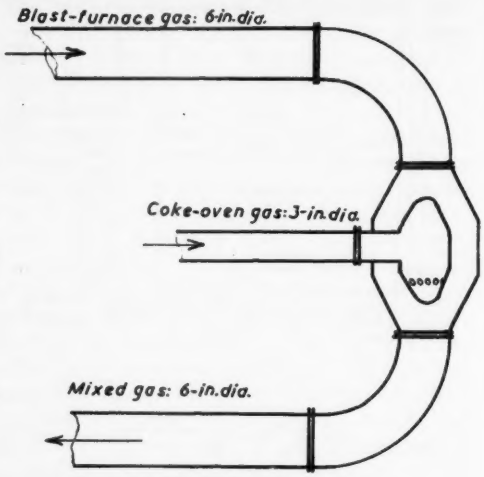

FIG. 4—Gas-mixing bulb

by fan, 11, to a stack, passing on their way a measuring orifice, 12, connected to a recording instrument and controlled by a damper. Since the waste gases do not pass through regenerators, some means of reducing their temperature before passing through the fan is necessary, and this is done by admitting the required volume of cold air by means of the by-pass, 13. The waste-gas fan is of sufficient capacity to handle this addi-

TABLE I—*List of Instruments and Range of Operating Conditions*

Instrument	Type	Experimental Range	Normal Operating Values
Gas meter	Kent ring balance (indicating) and recording)	2,000–10,000 standard cu. ft./hr.	6,000 standard cu. ft./hr.
Air meter	,, ,, ,,	2,000-30,000 standard cu. ft./hr.	9,000 standard cu. ft./hr.
Waste-gas meter	,, ,, ,,	According to air and gas flows and draught used.	
Ingoing air and gas temperatures	Spot test by mercury thermometers	15–20° C.	15–20° C.
Waste-gas temperatures ...	Suction pyrometer at furnace; ordinary thermocouple at measuring orifice	According to gas and air flows	700–800° C. 400–500° C.
Roof temperature	Base-metal couples in roof ...	Up to 1200° C.	400° C. ingoing end, to 1000° C., outgoing end.
Roof-pressure recorder	Electroflo diaphragm type ...	− 0·06 to + 0·06 in. W.G.	+ 0·02 in. W.G.
Water meter for main calorimeters	Kent U-tube (K.M.D. type) ...	Up to 600 gal./hr.	250 gal./hr. for low heat rates. 450 gal./hr. for high heat rates.
Temperature rise in calorimeter water	Nickel resistance thermometers	20–50° C.	30° C.
Calorific value of gas	Simmance recording calorimeter	100–200 B.Th.U./standard cu. ft.	150 and 180 B.Th.U./standard cu. ft.
Principal measuring orifices ...	Kent P-type in stainless steel ...	...	...

tional volume and is provided with water-cooled bearings.

The water supply to the main calorimeter tanks is provided from a high-level tank and passes through a 2-in. pipe (not shown in Fig. 2) provided with a measuring orifice and recording instrument. After passing through the calorimeter tanks in the hearth the water is run to waste.

All the pipework is constructed so as to afford the necessary straight lengths so that the locations of the measuring orifices conform to British Standard Specification No. 1042 for flow measurement.

Instruments

A list of the instruments installed and the range of operating conditions encountered in normal experimental work are given in Table I. The instruments are all well-known standard products, with the exception of the resistance thermometers used for the measurement of the heat transferred to the bath. For this measurement it was desired to obtain a differential instrument which would record a temperature rise of the order of 20–50° C. within $\frac{1}{10}$° C. This was done by connecting resistance thermometers

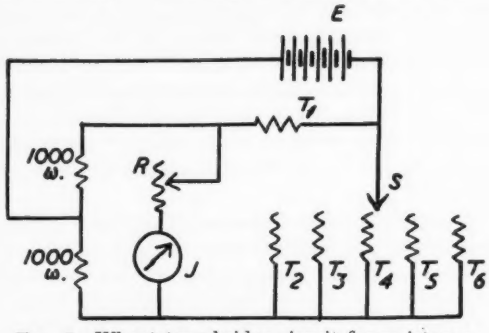

FIG. 5—Wheatstone-bridge circuit for resistance-thermometer connections

in a Wheatstone-bridge circuit as indicated in Fig. 5, making use of six resistance thermometers, each of 11·12 Ω resistance at 15° C. This is a conventional bridge circuit in which one arm is furnished by the thermometer T_1, which is inserted in the cold-water supply to the hearth tanks. The second arm is provided by any of the five " hot " thermometers T_2 to T_6. Four of these are inserted at the exits from the hearth tanks, (the difference $T_1 - T_5$ giving the total temperature rise in the bath), the remaining thermometer, T_6, being brought inside the instrument house for calibration purposes. Any thermometer could be brought into circuit by the switch S.

The remaining two arms of the bridge are provided by accurate 1000-Ω resistances, and the necessary e.m.f. is furnished by a 12-V. accumulator, E. The indicating instrument, J, is designed for temperature measurement from a base-metal thermocouple reading up to 1100° C. The various resistances are chosen so that the e.m.f. across J produced by a temperature

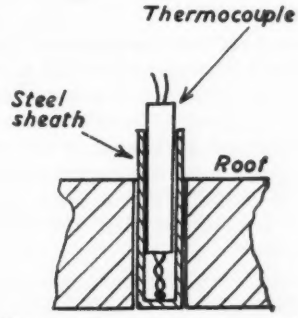

FIG. 6—Thermocouple assembly

difference of x° C. between T_1 and any thermometer T_2 to T_6 is the same as that produced by a base-metal thermocouple operating with a junction-temperature difference of $10x$° C. In this way the instrument J can be easily read to a temperature difference of 0·1° C.; for example, if the instrument J read, say, 336° C., the temperature rise in the water was 33·6° C.

For calibration purposes the thermometer T_6 is placed in boiling water, the temperature of which is read by a sensitive mercury thermometer in conjunction with the temperature read by a similar mercury thermometer immersed in the ingoing water. The variable resistance, R, is adjusted until the temperature difference read on the instrument J agrees with that shown by the mercury thermometer. This calibration, which is carried out several times daily, compensates for variations in the e.m.f. of the battery and in the various contact resistances. The other instruments are frequently calibrated according to the methods advised by the manufacturers.

The measurement of roof temperature offers a problem which at the time of writing is not completely solved. It is desired to measure the roof temperature in many places, and, owing to the small size of the openings in the furnace, interference from the flame, and the relatively low temperature of operation of the furnace, methods of radiation or photo-electric pyrometry cannot be applied. Recourse had to be made to thermocouples, and Fig. 6 shows the type of thermocouple assembly used. A closed-end steel sheath is inserted through a hole in the roof and the closed end ground flat so that it makes a practically continuous surface with the inside of

the furnace roof. A chromel–alumel couple is inserted in the sheath and the tip pressed firmly against this flat end. This appears to give a reliable reading of the temperature of the inner surface of the roof, as it agreed with the readings of a Foster mirror-type total-radiation pyrometer sighted on the sheath tip immediately after the gas was shut off.

The subsequent cooling curves of the radiation pyrometer and thermocouples did not follow the same slope, as the thermocouple was able to draw on stored heat in the inner part of the roof brick (*via* the conducting steel sheath), but as extrapolation of these cooling curves to zero time (the time of shutting-off the gas) showed agreement, it was assumed that the thermocouples read correctly.

Unfortunately, trouble was experienced with the steel sheaths and thermocouples burning-out after a few days in the hotter parts of the roof, where temperatures of over 1000° C. are attained, and specially robust couples and heat-resisting sheaths are now being installed.

Automatic damper control* is fitted to the furnace, but, as might have been anticipated before it was installed, it is found that the automatic-control gear has no work to do when the furnace is running. Owing to the close control of gas and air flow and the absence of variables in running conditions, for any given conditions the pressure in the furnace remains constant without alteration of the damper setting.

Accuracy of Measurement

It has already been stated that one of the reasons for abandoning any attempt to melt steel in the experimental furnace was the difficulty of taking measurements with any degree of accuracy at steelmaking temperatures. Work at lower temperatures brings reasonably accurate measurement within the bounds of practicability. At the same time, it must be remembered that work at 800–1000° C. on a semi-plant scale is in many ways not conducive to the same degree of accuracy as can be attained on a laboratory bench. Consequently it is necessary to consider further the accuracy obtainable, as the accuracy governs to a great extent the experimental programme.

The prime quantities which have to be measured are the heat put into the furnace and the heat leaving the furnace by various means (heat transfer to the hearth, wall losses, and so on). Measurements of experimental conditions (*e.g.*, roof temperature, furnace pressure, and atmospheric conditions) are required at the same time. The various meters which are installed are all subject to errors, and these will now be considered *seriatim*.

Gas Meter, Air Meter, and Waste-Gas Meter—In the gas, air, and waste-gas meters the error likely to be encountered with a ring balance meter and orifice system in an ideal location is about $\pm$ 2%, the error increasing somewhat at low rates of flow.

Gas Calorific Value—The maximum error in the recording of the calorific value of the gas is of the order of $\pm$ 2–3 B.Th.U./cu. ft., or say 2%.

Heat Input to Furnace—The heat input to the furnace was calculated by multiplying the gas volume recorded on the gas meter (reduced to standard conditions) by the gas calorific value as measured under similar standard conditions, that is $\pm$ 4%.

Water Meter—The error in the water-meter reading should not exceed 1%.

Temperature Rise in Calorimeter Water—Temperature differences in the calorimeter water can be easily read to $\pm$ 0·1° C. on the instrument used. The bridge set-up for measuring this temperature, and the method of calibration employed, automatically compensate for minor changes in battery e.m.f. or circuit resistance, and the main source of error in this reading is due to the observational error (0·1° C.). Since the temperature rise in the water is of the order of 30° C., this means an error of about $\pm$ 0·3% in the measurement of temperature rise.

Waste-Gas Temperature—At the time of writing, the waste-gas-temperature reading is the one subject to the greatest inaccuracy, as the suction pyrometer used is probably subject to an error of up to 5% (reading low), according to published data on such pyrometers.[8] In addition, leaks inevitably arise in the waste-gas system. Even with constant attention to patching and repairs it was found impossible to keep the waste-gas system tight. In theory, this should not matter as far as measurement of waste-gas heat loss is concerned, as infiltration of cold air after the working chamber, whilst giving an increased waste-gas volume, should also give a correspondingly reduced waste-gas temperature. In practice, the infiltration of cold air near the point of temperature measurement brings about stratification in the waste-gas flue, and the infiltrated cold air may not be "appreciated" by the suction pyrometer. The resulting calculated waste-gas loss is therefore high. This error, fortunately, tends to counterbalance the low-reading error fundamental to the suction pyrometer.

* Lent by courtesy of the Electroflo Meters Co., Ltd.

A summary of the errors in measurement which may be anticipated is given in Table II.

TABLE II—*Summary of Estimated Errors in Measurement*

Item	Error		
	Flow Measurement	Temperature or Calorific Value Measurement	Total
Heat input to furnace	± 2%	± 2%	± 4%
Heat transferred to hearth	± 1%	± 0·3%	± 1·3%
Heat in waste gases	± 3%	± 5%	± 8%

In determining the heat balance of the furnace there remains one further item, *viz.*, the surface heat losses. In most heat balances these are estimated by difference, but in the present series of investigations an attempt to assess them was made by a series of measurements of surface temperature by means of a contact pyrometer. These are discussed later.

Operation of the Furnace

At present the furnace is operated on a 24-hr. day basis, with three shifts. In the earlier stages of the work, shortage of technical staff necessitated discontinuous operation on a one- or two-shift basis. The size of the furnace made it advisable to adopt precautions during lighting-up similar to those adopted for full-size furnaces, and as gassing had to be carried out daily before 24-hr. operation was started, steps were taken to make this easy and safe by fixing a steam pipe to the ingoing gas system. Gassing was then easily carried out by (*a*) opening the furnace doors and starting the exhaust fan, (*b*) turning steam into the ingoing gas system until all the air had been driven out, and (*c*) placing a lighted flare against the gas port and turning the gas on and the steam off. Gassing was always carried out on blast-furnace gas alone, the coke-oven gas being added as soon as the furnace was operating satisfactorily.

From the time of gassing the furnace, about 5–6 hr. were required before conditions could be regarded as sufficiently steady for measurements to be taken. A careful check, making use of the data from over 500 experiments, showed no correlation between the percentage of heat transferred to the hearth and the time after gassing, for values of the latter between 6 and 90 hr., thus showing it was possible to bring the furnace up to effective working heat in 6 hr. The experimental procedure, once the steady state is reached, is as follows:

(i) Check calibrations of the resistance thermometers and gas calorimeter are made.

(ii) Readings of all four resistance thermometers are taken at 10-min. intervals until the amount of heat entering the bath has remained steady for at least 10 min. (this ensures the steady state having been reached).

The other instrument readings required are obtained from their respective continuous recordings, but a recording instrument has not, as yet, been installed to measure the temperature rise in the main calorimeter water, since the required accuracy of temperature measurement demands a visual observation taken at the time.

Having determined the heat transfer to the bath under one known set of conditions, the next set of conditions is then set up and maintained for about 1 hr. Except in special cases, 1–2 hr. are sufficient for a steady state to be reached after such a change in conditions. Nevertheless, a slight "creep" sometimes persists for several hours, but this "creep" is not sufficient to warrant the great loss of time which would result from waiting for the last traces to be eliminated on every occasion. This effect is exhibited as a kind of "hysteresis loop," which is obtained if a characteristic curve is prepared by gradually increasing the gas flow in successive experiments, followed by gradually returning to the original flow. Several of these "hysteresis loops" were obtained and were always of very small area; the scatter of the points was less than that occasioned by the other variables.

Although the variables in the experimental furnace are under much closer control than in a production furnace, the experimental points show some scatter. The proportion of heat transferred to the hearth is very sensitive to certain variables, notably gas calorific value and pressure in the working chamber. It has not been possible to eliminate chance variations of a few per cent in these under any set of controlled conditions, and this causes some scatter.

With the above considerations in mind, no difficulty is found in preparing curves from the collection of experimental points; if the conditions during any set of experiments give a greater scatter than usual, the number of experiments is appropriately increased. Figs. 7 and 8 are examples of small and large scatters in this respect, and it will be seen that even under difficult conditions it was possible to obtain curves of sufficient precision by increasing the

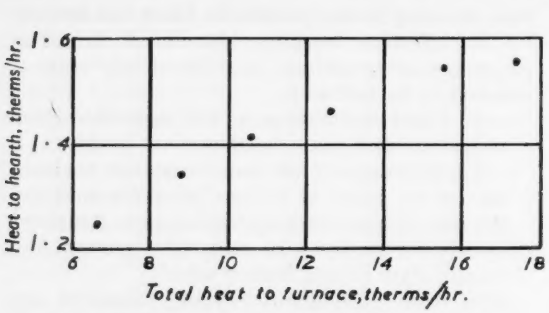

FIG. 7—Characteristic curve (small scatter)

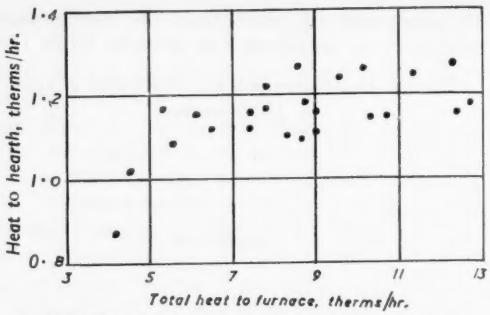

FIG. 8—Characteristic curve (large scatter)

number of experiments appropriately. It has to be borne in mind that small differences between curves which are drawn through a series of points showing scatter, as in Fig. 8, may not be significant, and, where necessary, a statistical significance test must be carried out before any importance is attached to a small difference between curves. Significance calculations on such curves are rather laborious, but after some experience it is usually easy to see when such calculations are advisable.

The layer of tiles covering the hearth tanks is frequently swept clear of dust, and the tiles themselves are removed, the tank surfaces swept, and the tiles replaced at weekly or fortnightly intervals. This procedure is necessary to achieve reproducible results. In early experiments a continuous fall in heat transfer was experienced over a period of weeks ; this fall was found to be due to a gradual but small increase of dust deposit on the hearth.

At frequent intervals, earlier experiments are repeated to ensure that results are reproducible. For this purpose one particular set of conditions is used as a " control," and a run under these conditions is made after each particular group of experiments is completed, to ensure that no external change has occurred.

Although one of the reasons for running at temperatures below the steelmaking level was the avoidance of the technical difficulties associated with melting, the operation of the furnace is by no means trouble-free. However, most of the troubles arise from external factors and are not inherent to the experimental furnace. The principal difficulties are :

(*a*) Shortage of gas because of burden problems at the blast-furnaces.

(*b*) Shortages of town's water for the calorimeters, necessitating " making-up " from the works' own supply, which tends to be undesirably warm.

(*c*) Making-up of the pipes to the gas calorimeter, owing to the sulphur content of the coke-oven gas.

Experimental work is thus occasionally held up whilst the above conditions are rectified.

The experimental results obtained are described in Part II, but an outline of some preliminary operating results is given herewith.

Heat Balance

Two heat balances were prepared based on runs lasting several hours, and a summary of the findings is given in Table IV.

TABLE IV—*Heat Balances*

Item No.		Test 1		Test 2	
		Heat Distribution		Heat Distribution	
		Therms/hr.	%	Therms/hr.	%
1	Heat to furnace	9·0	100·0	9·18	100·0
2	Heat usefully transferred to hearth	1·33	14·8	1·15	12·5
3	Heat in waste gases	5·47	60·8	5·91	64·4
4	Heat in water flowing through seals	0·81	9·0	0·59	6·4
5*a*	Radiation and unaccounted-for losses (by difference) ...	1·39	15·4	1·53	16·7
			100·0		100·0
5*b*	Radiation and unaccounted-for losses (calculated from surface temperature[6])	1·23	—	1·11	—

Item 5*a* is obtained by difference. As mentioned above, a rough check on this was obtained by taking numerous surface-temperature measurements and calculating the heat loss by radiation and convection from the customary formulæ.[6] These figures are in fair agreement with what is observed by difference, and show that there is no large unaccounted-for loss.

A comparison between these and figures for production furnaces[5] shows that, whilst the heat distribution is of the same order, the model-furnace radiation losses are lower (owing to the lower temperature of operation) and the proportion of heat lost in the waste gas is correspondingly higher.

The proportion of heat usefully transferred to the bath is lower than that encountered in practice, and this could have been increased to 25% or more by reducing the amount of insulating-tile covering on the hearth tanks. The efficiencies encountered in production furnaces depend largely on the amount of cold material melted, and a furnace employing a high proportion of cold metal will always show a high thermal efficiency (calculated on a conventional basis) on account of the large temperature difference between flame and charge which exists for a considerable part of the melting-down period. During the refining period, when the bath is "flat," conventional efficiencies are much lower and may drop to a few per cent or less, according to the method of calculation.

In view of the above factors, it can be said that the thermal efficiencies in the region of 10–20% obtained during the experiments represent production conditions with a bias towards the later stage of melting-down or the early stage of refining (*see* Appendix).

IV—Programme of Experiments

The programme for the experiments was divided into two main sections, namely, (1) the investigation of design factors and (2) the investigation of operation factors. For example, some of the factors investigated were :

Design Factors

(*a*) Effect of port size.
(*b*) Effect of port design.
(*c*) Effect of chamber design.

Operation Factors

(*d*) Effect of gas rate.
(*e*) Effect of air/gas ratio.
(*f*) Effect of calorific value of the gas.
(*g*) Effect of pressure in the working chamber (stack draught).

The procedure for any one port design was to alter the air- and gas-port sizes, and for each size to prepare a set of characteristic curves for four air/gas ratios (roughly 15% deficiency, theoretical air, 25% excess, and 55% excess). The effect of working-chamber pressure and gas calorific value was separately investigated for some, but not all, of the port designs and sizes ; it was not thought necessary for the effect of these variables to be examined for every port for which characteristic curves are obtained, as the general effect of calorific value and chamber pressure has so far been very similar for all ports.

In addition to the characteristic curves which were the main goal of the experiments, other curves connecting heat transfer to the hearth with furnace pressure and gas calorific value, and analysing the distribution of heat transfer in various parts of the hearth, were prepared where required.

The furnace has now been operated successfully for a period of over two years, at first with a block of more or less conventional design and latterly with various forms of the Maerz block. The results of the experiments with the conventional block are described in Part II ; later experimental results will be given in following Parts.

V—Acknowledgment

The authors desire to record their thanks to the Management of the Shelton Iron, Steel and Coal Co., Ltd., for the experimental facilities provided.

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APPENDIX—*Further Notes on Similarity between Experimental and Production Furnaces*

Aerodynamic Similarity—Reynolds Numbers in Gas Port

The Reynolds numbers for an 80-ton production furnace and for the experimental furnace are calculated as in Table V.

The quantity ν is calculated from the data in British Standards Specification 1042, p. 29, and v and d for production furnaces are calculated from the data given in a symposium on steel-making organized by the Iron and Steel Institute.* High Reynolds numbers are generally associated with high velocities and small port diameters, since change of port diameter affects v more than d in the formula.

From Table V it will be seen that the range of Reynolds numbers is of the same order in both production and experimental furnaces; that in production furnaces is slightly higher. There is, however, overlap over a considerable range, and in any case Re in both furnaces is so well above the turbulent limit (Re = approximately 2,000) that the similarity may be regarded as satisfactory.

Thermal Similarity

The similarity criterion suggested by Thring (*see* p. 394) is the dimensionless group:

$$\frac{Q}{A\,\sigma\,E\,(T_1^4 - T_2^4)},$$

where Q is the heat input per hour,
A is the area of the heat-absorbing surface,
σ is Stefan's constant,
T_1 is the absolute temperature of the heat-emitting surface,
T_2 is the absolute temperature of the heat-absorbing surface,
and E represents the "combined emissivity" of the surface,
all in self-consistent units.

* *The Iron and Steel Institute*, 1938, *Special Report No.* 22, p. 379.

TABLE V—*Calculation of Reynolds Numbers* (Re) *for 80-ton Production Furnace and Experimental Furnace*

	High Reynolds Conditions				Low Reynolds Conditions			
	v, cm./sec.	d cm.	ν c.g.s. units	$Re = \frac{vd}{\nu}$	v, cm./sec.	d cm.	ν c.g.s. units	$Re = \frac{vd}{\nu}$
Production furnace (temperature of gas in port, 1100° C.)... ...	4000	46	1·58	116,000	1500	50	1·58	47,500
Experimental furnace (temperature of gas in port, 20° C.)	1350	8	0·159	68,000	470	12	0·159	35,500

TABLE VI—*Calculation of "Thring Numbers" for 80-ton Production Furnace and Experimental Furnace*

Quantity	Production Furnace		Experimental Furnace	
	High Number	Low Number	High Number	Low Number
Q (B.Th. U./hr.)	600×10^5	400×10^5	16×10^5	6×10^5
A (hearth area in sq. ft.)	300	500	18·3	18·3
Estimated flame temperature, T_1 (°F., absolute)	3800	3800	2700	2700
Estimated hearth surface temperature, T_2 (°F., absolute)	3200	3200	900	900
Approximate "Thring number" $= \frac{Q}{A\,\sigma\,E\,(T_1^4 - T_2^4)}$	2·22	0·88	1·93	0·72

Note: $\sigma = 17{\cdot}3 \times 10^{-10}$ B.Th.U./sq. ft./hr. E is taken as 0·5 for both furnaces

The use of this formula assumes that the main heat transfer is by radiation; this is justified when dealing with high-temperature furnaces.

The "Thring numbers" for the existing experimental furnace and an 80-ton production furnace (on which the experimental furnace is based) are roughly calculated in Table VI.

Thus the ranges of "Thring numbers" for the production and experimental furnaces are in satisfactory agreement, and it may be said that thermal conditions in the model are sufficiently near to similarity with practice, despite the difference in temperature. It will be noted that the attainment of similarity depends on the relatively high value for T_2 in a production furnace (1500° C.) and thus the experimental furnace conditions correspond to a production furnace with the bath almost flat or melted (*see* p. 403).

Part II—THE EFFECTS OF GAS RATE, PORT SIZE, AIR/GAS RATIO, FURNACE PRESSURE, AND GAS CALORIFIC VALUE

By A. H. Leckie, Ph.D., J. F. Allen, B.Sc.,* and G. Fenton, B.Sc.**

SYNOPSIS

The effects of gas rate, port size, air/gas ratio, furnace pressure, and gas calorific value in the experimental furnace described in Part I have been studied. In general, the best results are obtained with the theoretical air/gas ratio, high furnace pressures, and high calorific value, provided that the port design is satisfactory. Some of the factors governing correct port design, in particular the relative areas of air and gas ports, are investigated and discussed.

I—INTRODUCTION

In the description of the experimental furnace (*see* Part I), it was pointed out that the normal experimental procedure would be the determination of a "characteristic curve" connecting the heat usefully transferred to the hearth with the total heat entering the working chamber. The general form and theory of these curves have already been described,[1, 2] and for brevity in the following discussion the term *C*-curve will be used to denote characteristic curves.

The investigations to be reported in this and in following Parts are concerned with variables of both furnace operation and furnace design. In this paper, the following factors are discussed, with particular reference to the proportion of the total heat entering the furnace which is transferred to the hearth:

Furnace Operation

(*a*) Gas rate.
(*b*) Air/gas ratio.
(*c*) Pressure in the working chamber (*i.e.*, draught).
(*d*) Gas calorific value.

Furnace Design

(*e*) Gas-port size.
(*f*) Air-port size.

Series of *C*-curves have been obtained for several conditions of the above variables, and the results used as a guide to the best practice and correlated with production-furnace operation as far as possible.

II—FURNACE OPERATION

(a) *Gas Rate*

The effect of variations in the amount of gas entering a furnace is very important, though the general absence of gas-metering equipment until recent years has made the study of such variations on production plant largely a matter of conjecture. With the aid of statistical methods,[1, 5] it has been possible to investigate the effect of gas rate on furnaces which have been fitted with gas meters, but even so, under production conditions, investigators have not been able to study the effect over a wide range without interfering with the steel-making process.

In the experimental furnace it has been possible to explore a range of gas flows varying from what, under production conditions, would be insufficient gas to operate the furnace to a flow which—assuming that combustion was satisfactory—would cause damage to the refractories.

Since the basis for the *C*-curves used for most of the investigations is the plot of heat transferred to the hearth against heat input as gas supplied, the effect of gas rate is automatically investigated

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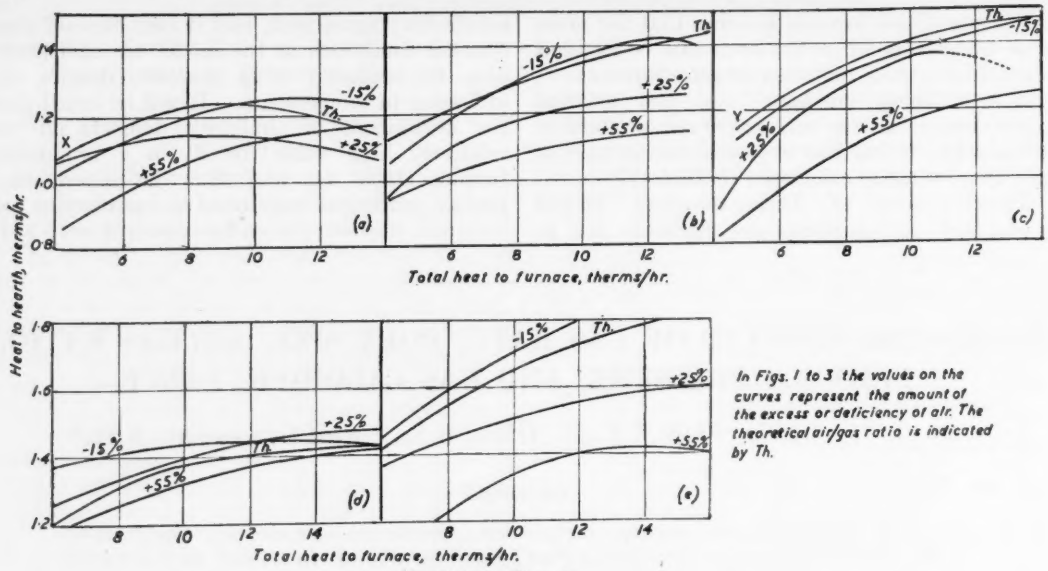
* British Iron and Steel Research Association.



Air-port area, 170 sq. in.
Gas calorific value :
(*a* to *c*). 150 B.Th.U./standard cu. ft.
(*d* and *e*), 180 B.Th.U./standard cu. ft.
Roof pressure,+ 0·02 in. W.G.
Port design No. 1.

FIG. 1—Effect of air/gas ratio : Large air port with (*a* and *d*) 26-sq. in. gas port, (*b*) 13½-sq. in. gas port, and (*c* and *e*) 9-sq. in. gas port

in most experiments. As the gas rate increases, the heat transfer to the bath increases correspondingly at first, but whilst under some conditions this increase continues throughout the experimental range, under other conditions once the gas rate exceeds a certain value the heat transfer to the hearth actually decreases with increasing gas flow. This is owing to the quantity of gas entering the furnace exceeding that which can be burnt in the furnace working chamber, so that the unburnt gas leaving by the exit port takes heat with it in the same way as excess air. A *C*-curve which exhibits such a drop is, therefore, indicative of either bad furnace operation (use of excess gas or insufficient air) or of a faulty port design which does not give rapid enough mixing for combustion to take place within the working chamber. A theoretical analysis of such curves has been given by Thring and Reber.[2]

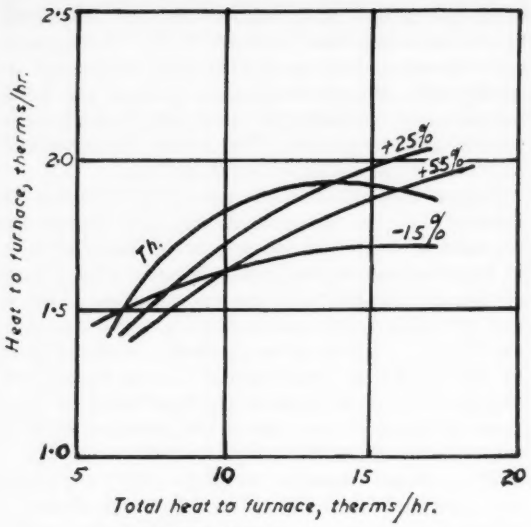
Curves *X* and *Y* in Fig. 1 are examples of the two types of curvature mentioned above and illustrate the advantages of a small gas port. Further discussion of the effect of gas rate is not necessary at this stage, as no abnormal results were obtained in any of the experiments. There was, however, ample confirmation of the finding (also observed in a few production furnaces) that under certain conditions excessive gas supply can lead to reduced heat transfer to the hearth.

(b) *Air/Gas Ratio*

The establishment of a correct air/gas ratio for furnace operation has long been a matter of interest. The chief obstacles to its investigation in production furnaces are that (1) accurate measurement of the quantity of air entering the furnace working chamber is not possible owing to infiltration between the admission valve (where the air can be metered) and the working chamber, and (2) any effect on furnace performance is inextricably lost in the effect of other uncontrolled variables.

Attempts have been made to study the effect of air/gas ratio in production furnaces statistically, and earlier calculations reported by one of the present authors[1] indicated that the high air/gas ratios, namely, in the region of 80% excess air, gave the highest thermal efficiencies. Subsequent statistical work failed to confirm this, and it has been found that on the furnaces from which the original conclusions were obtained the high air/gas ratios were always associated with a low gas rate, *i.e.*, the air rate was not altered widely but the gas rate did vary. Now the theory of *C*-curves for open-hearth furnaces shows that thermal efficiency tends to fall as the gas rate increases,* and consequently the highest thermal

* This is for open-hearth furnaces. Whether the curve rises or falls depends on the nature of the heat-transfer plant and process (*see* reference 2).



Gas-port area, 26 sq. in.
Air-port area, 25 sq. in.
Gas calorific value, 180 B.Th.U./standard cu. ft.
Roof pressure, + 0·02 in. W.G.
Port design No. 2.

FIG. 2—Effect of air/gas ratio : Small air port

efficiencies are generally associated with the lowest gas rates. It should be emphasized that a high thermal efficiency in an open-hearth furnace does not necessarily correspond to a maximum output of steel. In general, the maximum production rate is reached at a point on the *C*-curve well above the point of highest thermal efficiency. This led, in the case of the two furnaces previously investigated,[1] to the false conclusion that a high air/gas ratio led to a high thermal efficiency.

Until recently, further investigations of the effect of the air/gas ratio on production furnaces failed to show any definite connection between the air/gas ratio as metered at the gas- and air-admission valves and the performance of the furnace. This may be due to the before-mentioned difficulties of measuring air and gas flow to the working chamber and to the swamping effect of the many other variables, or it may be a correct conclusion that substantial changes in the air/gas ratio do not, in fact, appreciably alter furnace performance. The recent work of Chesters and Thring and their colleagues[5] has indicated, however, that the air/gas ratio has an effect on heat transfer to the hearth, although these investigators again encountered the difficulty of measuring the true air/gas ratio, as distinct from that measured at the valves.

It was therefore important to carry out air/gas ratio experiments on the experimental furnace, on which the flow can be metered accurately, the performance assessed with some precision, and the variables closely controlled. It is appreciated that conclusions drawn from results obtained with the experimental furnace, operating at temperatures of 800–1000° C., require careful consideration in relation to production conditions, but in view of the satisfactory similarity relationships between the experimental furnace and production furnaces—discussed in Part I—it may be expected that experiments carried out at lower temperatures will throw much light on the effect of the air/gas ratio on combustion conditions in open-hearth-type ports.

A series of experiments under various conditions was carried out with (*a*) 15% deficiency of air (*b*) theoretical air, (*c*) about 25% excess air, and (*d*) about 55% excess air. In all these experiments the heat transferred to the hearth was plotted against the total heat supplied to the furnace.

Figures 1*a* to 1*e* show the results of these experiments with port design No. 1 (*see* Fig. 20). The heat transferred to the hearth of the furnace at various gas rates and air/gas ratios is shown. Fig. 2 is a set of *C*-curves for a different design of port (port design No. 2 in Fig. 20). Over much of the range the same effect of excess air on efficiency is seen, but the effect of excess air on eliminating the falling portion of the curve at high gas rates is much more marked. The explanation of this is later discussed in the consideration of port design.

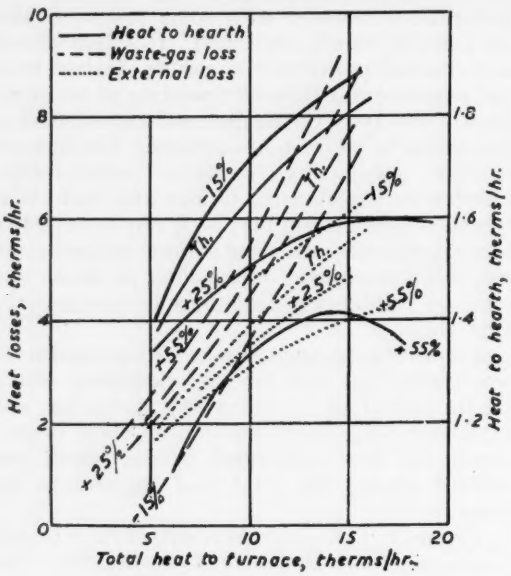
In all the *C*-curves shown in these diagrams no significance should be attached to small differences between the curves, as these curves were prepared from an array of experimental points showing some scatter. Generally speaking, differences of the order of 15% or more can be regarded as significant (*e.g.*, 0·15 therm/hr. in a heat transfer of 1 therm/hr.).

From Figs. 1 and 2 it may be seen that :

(*a*) In general, excess air has the expected effect of reducing the heat transferred to the bath.

(*b*) A 15% deficiency of air has little effect. In the diagrams the 15% deficiency curve shows sometimes better results and sometimes worse results than theoretical air, but the difference is so small that, in view of the scatter of the points from which these curves were prepared, no real significance can be attached to it.

(*c*) The bad effect of excess air is emphasized as the gas port is made smaller. In other words, decreasing the size of the gas port leads to improved furnace operation where a large excess of air is not used. With the large gas port the effect of air/gas ratio was less marked,



Gas-port area, 9 sq. in.
Air-port area, 170 sq. in.
Gas calorific value, 180 B.Th.U./standard cu. ft.
Roof pressure, + 0·02 in. W.G.
Port design No. 1.

FIG. 3—Effect of air/gas ratio on heat balance

particularly with gas of high calorific value. The effects of gas-port size are discussed later.

(*d*) Some of the curves with the largest gas port do exhibit the phenomenon of a drop in heat transfer to the hearth once a certain heat input is exceeded, but the effect of excess air is to eliminate this. For instance, in Fig. 1*a* all the curves except that for 55% excess air show this maximum, with the result that at the highest heat inputs (gas flow) 55% excess air gives better results than theoretical air. This effect is more pronounced in Fig. 2, where 25% excess air is sufficient to show it. These points are discussed later in this paper.

It is generally assumed that excess air is undesirable because it increases the waste-gas heat losses from the working chamber. Since waste-gas volume and temperature are both measured in the experimental furnace, it is possible to check whether this is indeed the case, or whether excess air has some other effect on combustion which accounts for its effect on heat transfer to the bath.

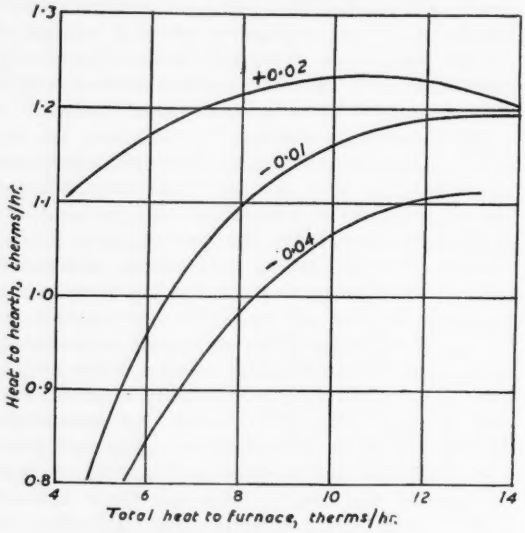
Unfortunately it is not yet possible to measure the heat content of the waste gases with great precision (*see* Part I), and most of the differences between the curves for various air/gas ratios in Fig. 1 are so small that they are in many cases absorbed in the scatter of the points when waste-gas losses are plotted against heat supplied to the furnace. However, in the case of Fig. 1*e* the effect of air/gas ratio is more marked and the difference in the measured heat content of the waste gases with theoretical air and with 55% excess air is significant. Figure 3 shows a plot of the heat balance over a range of heat inputs and with various air/gas ratios. The curves for external heat losses are obtained by difference.

Figure 3 shows that the use of 55% excess air instead of the theoretical amount causes an increase in waste-gas losses greater than the drop in heat transfer to the bath. That is, with excess air the external heat losses are reduced. This is not surprising, as the excess air reduces the overall temperature of the furnace. At the middle of the range of heat inputs the average roof temperature at a position corresponding to that over the tap-hole was about 50° lower with 55% excess air than with theoretical air.

The above results confirm the generally accepted view that the reduced furnace efficiency brought about by excess air is mainly due to the cooling effect.

(c) *Pressure in Working Chamber*

In recent years increasing attention has been given to control of the pressure in the melting chamber of furnaces. Although the pressure under the roof is less than 0·1 in W.G., changes of 0·01 in. are quite sufficient to cause great fluctuations in the quantity of cold air infiltrated through the doors and other openings in the



Gas-port area, 26 sq. in.
Air-port area, 170 sq. in.
Air/gas ratio, theoretical.
Gas calorific value, 150 B.Th.U./standard cu. ft.
Port design No. 1.

FIG. 4—Characteristic curves showing effect of furnace pressure. The values on the curves represent the roof pressure, in inches water gauge

furnace. If this pressure is too low there is appreciable air infiltration, whilst if it is too high a large quantity of flame and hot gases issue from the various openings, causing damage to the furnace structure and much apparent loss of heat. However, it has been shown theoretically[3] that flame "sting" is not serious compared with air infiltration and that relatively high operating pressures are desirable.

Whilst many operators have found that this argument is confirmed in practice, others do not agree, and the effect of pressure changes is obviously a subject for investigation on the experimental furnace.

In this series of experiments a set of C-curves for various pressures was obtained, and experiments have also been carried out in which all conditions except pressure have been kept constant, the variation in heat transfer being measured.

Figure 4 shows C-curves for three different roof pressures, other conditions being constant. The proportion of heat transferred to the hearth decreases as the pressure decreases throughout the range, but at the highest heat inputs the curve for $-0{\cdot}01$-in. W.G. pressure approaches that for $0{\cdot}02$ in. The conditions selected for this experiment were those in which the $0{\cdot}02$-in. pressure curve showed a maximum (*i.e.*, at the highest heat inputs combustion was incomplete), in order to observe the effect of pressure on this. It is suggested that the additional infiltrated air at the lower pressures assists combustion at the highest gas flow, thus tending to eliminate the maximum. At the same time this infiltrated air, although perhaps aiding combustion where it would otherwise be inefficient, has the general effect (*cf.* excess air, already discussed) of reducing the proportion of heat transferred to the bath.

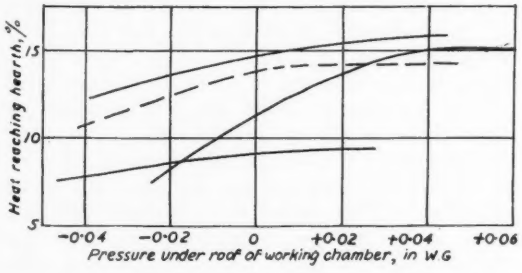


Fig. 5—Effect of furnace pressure. Four curves, each representative of a different set of constant operating conditions, some of which give relatively high, and others relatively low, efficiencies. Under all conditions, increasing the pressure leads to improved results

Figure 5 shows a series of curves for which pressure is the only variable. In all the curves the efficiency of the furnace as regards heat transfer to the bath increases with roof pressure, up to a point where the efficiency seems to remain roughly constant with further increase of pressure. Theory[3] shows that the minimum net heat loss due to air infiltration and external-flame loss is likely to occur where the atmospheric-pressure line in the furnace is at or about the level of the bottom of the doors. Now the pressure at the roof is given by :

$$p = 0{\cdot}0147\, H\, (1 - \frac{T_a}{T_f}),$$

where :

p = pressure under roof (inches water gauge),
H = height of roof above atmospheric-pressure line (feet),
T_a = atmospheric temperature (degrees Kelvin),
T_f = furnace temperature (degrees Kelvin).

In the experimental furnace $H = 2$ ft. $1\frac{1}{2}$ in., T_a is assumed to be 15° C. (288° K.) and T_f to be 800–1000° C. (1073–1273° K.) This leads to a value of $+\ 0{\cdot}025$ to $+\ 0{\cdot}03$ in. W.G. for the point at which further increase of pressure does not lead to increased efficiency.*

* In a production furnace where H = say, 8 ft. and $T_f = 1700°$ C. (1973°K.), p is calculated as $0{\cdot}1$ in., a value which is seldom, if ever, attained in practice. If $H = 6$ ft., p is $0{\cdot}075$ in.

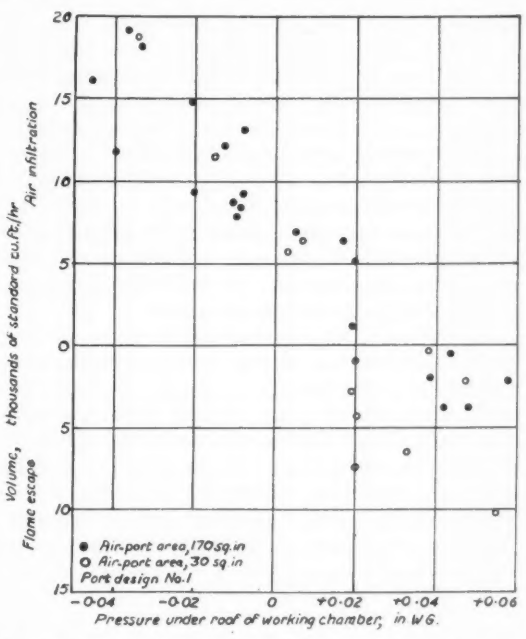


FIG. 6—Effect of furnace pressure on infiltration and flame escape

In Fig. 5 the point at which most of the curves become level is slightly above 0·03 in. W.G. and in fair agreement with theory. The exception to this is the curve for the small air port, which becomes level at a measured roof pressure almost imperceptibly above atmospheric. The reason for this is that with this air port the high-velocity air impinging on the gas stream tends to split the flame and force part of it against the doors. It was observed that with this port the flame sting seemed much greater than usual, and this was undoubtedly owing to the flame being blown against the doors by the high-velocity air entering above the gas. This would have the same effect as a roof pressure much higher than the measured value in keeping out infiltration, and it is not surprising that the point of maximum efficiency should be reached at a lower value of the measured roof pressure in the case of the small air port. In practice, splitting of the flame in this way would cause severe wear of the front and back linings, and it is preferable to reduce air infiltration at the doors by uniform increase of furnace pressure rather than by blowing the flame against the door.

Figures 4 and 5 emphasize the desirability of as high a pressure in the working chamber as is practicable. It has already been mentioned that translation of the results on to a production-furnace basis leads one to recommend a roof pressure in the region of 0·1 in. W.G. Such a pressure would cause excessive flame sting in most furnaces, and is higher than is generally regarded as practicable. The conclusion from these experiments (which had already been reached on theoretical grounds) is that furnaces should be operated with the highest roof pressure that is practicable. Thus, door arches and jambs should be water-cooled,* not only to reduce the gap between door and furnace body, but to permit of appreciable emergence of flame without damage to the structure. Door-lifting gear should also be designed with this in view.

The substantial effect of furnace pressure on infiltration is shown in Fig. 6. The volume of air infiltrated at the furnace openings, or the volume of flame and hot gases leaving these openings, may be roughly determined from the difference between the calculated waste-gas volume and the metered waste-gas volume. In Fig. 6 this difference is plotted against furnace pressure. There is, of course, considerable scatter, as the metered waste-gas volume varies according to the state of repair of the furnace, but the correlation for the large air port is very strong. The effect of the small air port in pushing the flame against the furnace doors is shown by the increased volume of flame sting with the small air port at high furnace pressures. At lower pressures the draught is sufficiently powerful to draw the flame away from the doors.

The above recommendations are quite independent of the obvious desirability of keeping all doors as well maintained and tight fitting as possible, whatever the working pressure.

(d) *Gas Calorific Value*

The desirability of using rich gas for the firing of open-hearth furnaces is not in serious dispute, although it has always been felt that high calorific value alone is not a sufficient criterion and that flame luminosity is equally important. It might be desirable to use a gas of somewhat lower calorific value if thereby a more luminous flame were obtained.

Generally speaking, the use of high-calorific-value gas has the advantages that a higher flame temperature is attained, thereby increasing the rate of heat transfer, and that the volume of waste gases per unit of heat input and the thermal losses arising therefrom are diminished.

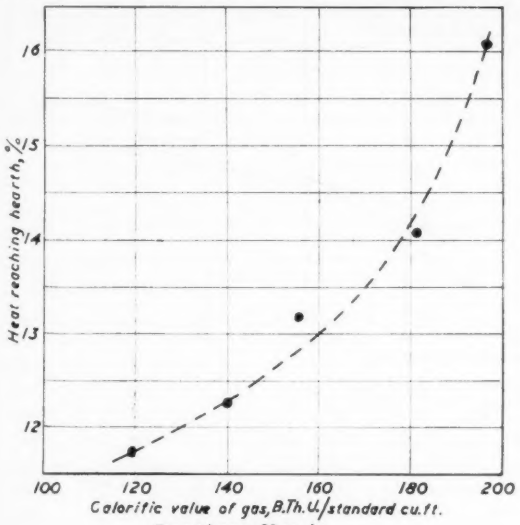
The low-temperature experimental furnace used for the present work is not suitable for an intensive study of these matters, as the principal variables concerned in an investigation of gas calorific value are flame temperature and luminosity,* both of which demand experimental temperatures as high as those reached in practice. Nevertheless, a few experiments in which the calorific value of the gas was made the principal variable are reported here, as the results are quite informative if it is remembered that they are purely qualitative.

In these experiments the heat input was maintained at about 9·75 therms/hr., this heat input being provided by a relatively low quantity of high-calorific-value gas or a relatively high quantity of low-calorific-value gas. The results of these experiments are shown in Fig. 7, in which the percentage of the heat input transferred to the hearth is plotted against the calorific value of the gas used. This shows the increased efficiency obtained by the use of high-calorific-value gas, but although the points appear to lie on a curve indicating a more rapid increase of efficiency at

* Water-cooling should, however, be carried out with moderation, so as to minimize the inevitable heat loss in such cooling; this heat loss, however, is generally less than the quantity of heat saved through reduced infiltration.

* The term luminosity here refers to the total emissive power and not merely that for visible radiation. A flame of apparent low luminosity may in fact have quite a high emissivity, owing to radiation in the non-visible region.

high calorific values, such a curvature would depend on the accuracy of the points at the extreme limits of the curve, and in the circumstances all that can be said is that the trend is upwards. This observation agrees with theory, but the extent to which it may be applied to production furnaces is doubtful. In practice the indications of Fig. 7 are not very closely followed, as we do not find that coke-oven-gas-fired furnaces have an overwhelming advantage in performance over producer-gas-fired furnaces. Fig. 8 shows



Gas-port area, 26 sq. in.
Air-port area, 170 sq. in.
Roof pressure, + 0·02 in. W.G.
Air/gas ratio, theoretical
Heat input, 9·75 therms/hr.
Port design No. 1

FIG. 7—Effect of gas calorific value. Other conditions constant

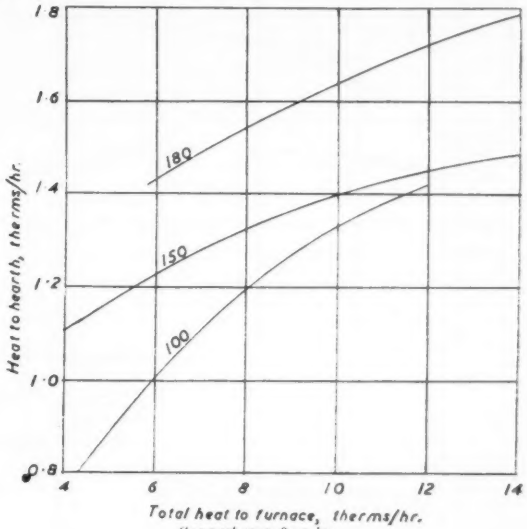
C-curves for three calorific values, and exhibits a similar trend.

(e) *Gas-Port Size*

The problem of port design is a favourite topic for discussion among furnace operators and has in recent years been the subject of an increasing amount of attention. A review of various suggestions for port design which have appeared in the literature from time to time is given by Chesters and Thring in their report[5] on an extensive and elaborate series of tests which were themselves a study of the behaviour in practice of three port designs.

The experimental furnace allows of a much wider range of experiments than can be carried out on production furnaces, but the permutations and combinations of port sizes, angles, and shapes which have to be tried in order to make a systematic investigation of the subject are very numerous and demand a long period of time. So far, three gas-port sizes and three air-port sizes have been investigated with one angle setting (port design No. 1) and a few experiments have also been carried out with a different layout of the small air port (port design No. 2). Port designs Nos. 1 and 2 are illustrated in Fig. 20.

Figures 9 to 11 show the effect of gas-port size; the curves are re-plots of the curves of Figs. 1 and 2. Figures 9 and 10 show that with a relatively large air port the smaller gas-port areas almost always give better heat transfer to the bath than larger



Gas-port area, 9 sq. in.
Air-port area, 170 sq. in.
Roof pressure, + 0·02 in. W.G.
Air/gas ratio, theoretical.

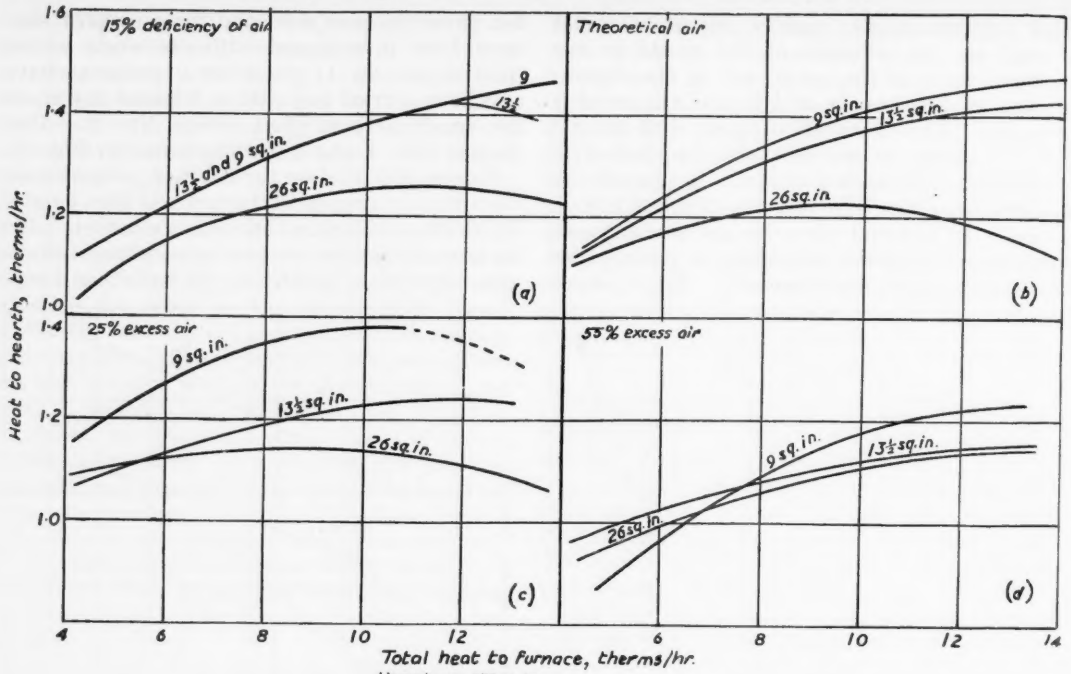
FIG. 8—Characteristic curves showing effect of gas calorific value. The values on the curves represent British thermal units per standard cubic foot

gas ports. The only exception is where a large excess of air is used; the difference between the curves is then either much diminished or becomes insignificant, and at low gas flows there is a suggestion that the small gas port loses its advantage. The position of the curve is reversed when the air port is small, as shown in Fig. 11. Here the best results are obtained with a large gas port.

Discussion of these results is left until the results of experiments on air-port size have been reported.

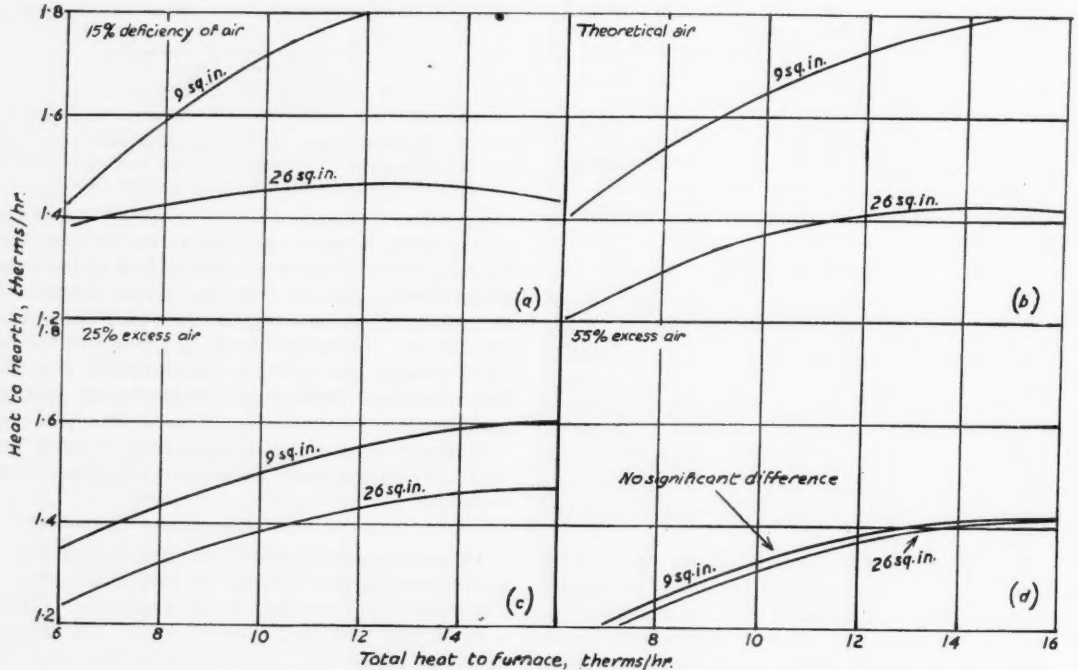
(f) *Air-Port Size*

Figures 12 and 13 show the results of experiments made with air-port size as the principal variable. The results at first sight seem conflicting. With a small gas port and theoretical air, the largest air port gives the highest efficiencies and the intermediate-sized air port gives the lowest, whilst the smallest air ports give efficiencies in



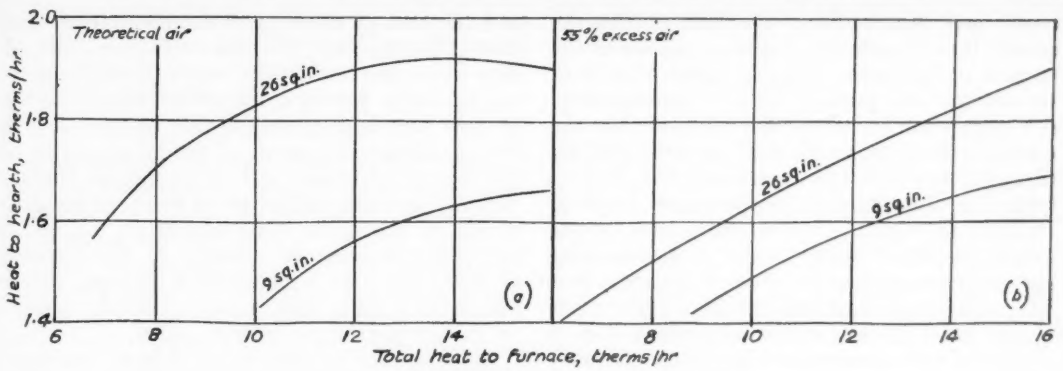
Air-port area, 170 sq. in.
Gas calorific value, 150 B.Th.U./standard cu. ft.
Roof pressure + 0·02 in. W.G.
Port design No. 1.

FIG. 9—Effect of gas-port size



Air-port area, 170 sq. in.
Gas calorific value, 180 B.Th.U./standard cu. ft.
Roof pressure, + 0·02 in. W.G.
Port design No. 1.

FIG. 10—Effect of gas-port size



Air-port area, 25 sq. in.
Gas calorific value, 180 B.Th.U./standard cu. ft.
Roof pressure, + 0·02 in. W.G.
Port design No. 2.

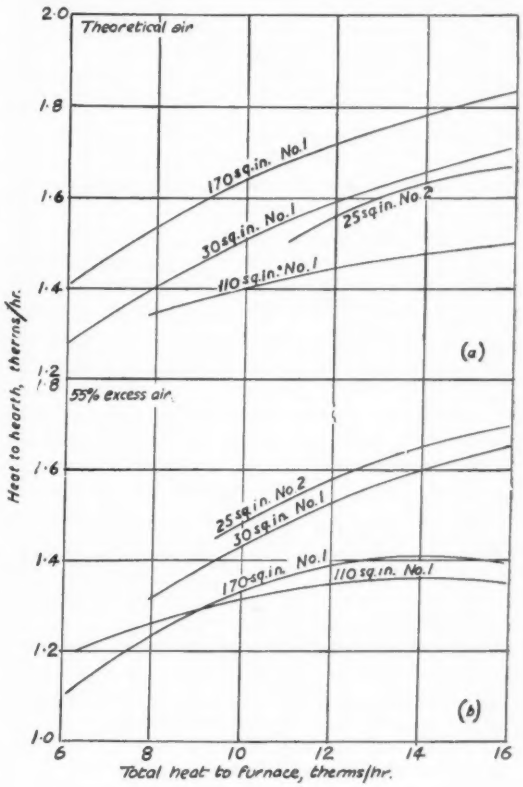
FIG. 11—Effect of gas-port size. Note inversion of curves compared with Figs. 9 and 10, owing to small air port

between the other two curves. When the air rate is increased to 55% excess, there is no significant difference between the large and intermediate air ports, but an advantage is shown by the small air ports. With a large gas port, experiments were not carried out with the 110-sq. in. air port, but the small air ports of 30 and 25 sq. in. give, on the whole, substantially higher efficiencies than the 170-sq. in. port.

The only serious anomaly occurs in the upper part of Fig. 13, where the curve for the 25-sq. in. air port is higher and that for the 30-sq. in. air port lower than what would be expected.* Apart from this, the behaviour of the curves in Figs. 12 and 13 is explicable on the assumptions that there is, for a given port design, an air/gas velocity ratio which gives a minimum efficiency, and that better combustion is obtained on each side of this.† Unfortunately, with the limited staff available, it was not possible to take continuous gas analyses to determine combustion efficiency. Waste-gas-volume and temperature measurement, or CO_2 determinations, do not give satisfactory explanations of the differences between the curves, for the following reasons :

(i) With the present apparatus the measurement of the heat content of the waste gases is not sufficiently accurate to show the differences of below 0·50 therms/hr. exhibited by the curves.

(ii) Air infiltration is likely to vitiate the measurements completely. Although the furnace pressure may be maintained at the standard + 0·02 in. W.G. on the recording instrument, it is observed that with a small air port the external flame leaving the doors and other furnace openings is abnormally large. This is owing to the high-velocity air splitting the flame and forcing it against the



Gas-port area, 9 sq. in.
Gas calorific value, 180 B.Th.U./standard cu. ft.
Roof pressure, +0·02 in. W.G.

FIG. 12—Effect of air-port size

* It has not yet been found possible to explain the large difference between the curves for the 25-sq. in. and the 30-sq. in. air port in Fig. 13*a*.

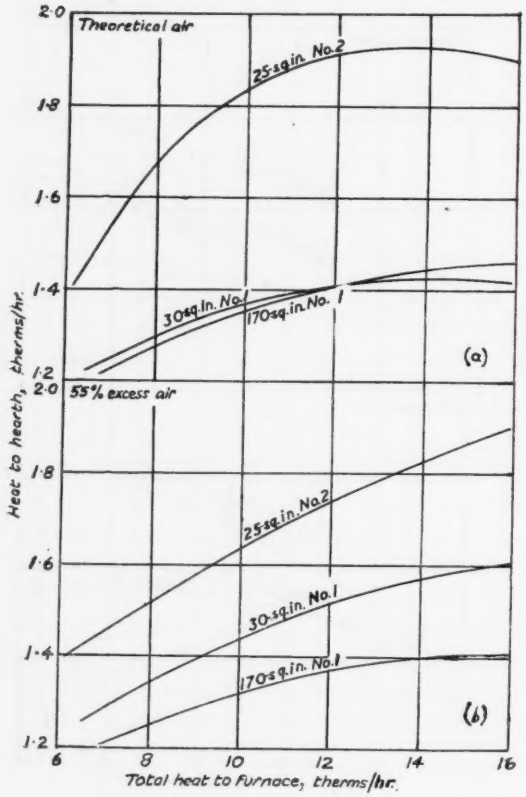
† *See* p. 416 for an extension of this view.

front and back walls of the furnace, with the result that the effective furnace pressure in the region of the door is much higher than that recorded on the pressure meter. Consequently, air infiltration at the doors is reduced. On the other hand, owing to the small air port, a higher draught between the stack and the furnace chamber is required to maintain the standard furnace pressure. Thus air infiltration to the waste-gas line through cracks in the outgoing gas system is increased. The net result is that the volume and heat content of the waste gas is no guide to combustion conditions. It is not clear how much of the increased efficiency attained with a small air port is attributable to the abnormal pressure conditions at the front wall, and how much to possibly improved combustion conditions brought about by high-velocity air.

The experimental observations obtained to date may be summarized as follows:

(1) When a conventional type of port is used, and the area of the air port is much larger than that of the gas port, the best results are obtained with an air/gas ratio in the region of the theoretical, excess air having a deleterious effect. As the air-port area is decreased to that of the gas port the advantage of control of the air/gas ratio to the theoretical is lost, and, under special circumstances—generally indicative of bad port design—excess air may show some advantage. Without exception, increase of pressure in the working chamber leads to a higher heat transfer to the hearth for a given heat input. This also applies to an increase in gas calorific value.

(2) With the normal type of large overhead air port the best results are obtained with a small gas port; very good results can also be obtained by the use of a very small air port (which in practice would need high-pressure air) in conjunction with a larger gas port. The use of an air port of medium area does not give good results.



Gas-port area, 26 sq. in.
Gas calorific value, 180 B.Th.U./standard cu. ft.
Roof pressure, + 0·02 in. W.G.

FIG. 13—Effect of air-port size

GENERAL DISCUSSION OF RESULTS

Air/Gas Ratio

Elementary theory has always indicated that excess air is undesirable, since this excess air simply removes heat from the furnace chamber. On the other hand, most melting-furnace operators consider that a certain excess of air, often in the region of 15–20%, is necessary to ensure efficient combustion. Two possible explanations of this are that excess air is really needed to burn the gas efficiently with some designs of port (which is in agreement with the conclusions of Rummel[4]) or that the excess air is required to maintain a layer of relatively cool air between the flame and the roof so that a high gas rate can be carried (*vide* Chesters and Thring[5]). It is evident that the port design of many furnaces is such that the correct volume of gas cannot be burnt without excess air, but if it were possible to design a port which did not require this excess air, much greater efficiencies would be obtained.

To sum up, the experiments on air/gas ratio have given a clear explanation of the wide divergence of opinion on this topic. The correct air/gas ratio for any furnace depends on the port design and the quantity of gas used, and it is not possible to specify a correct ratio for general use in furnaces, although any necessity to depart greatly from the theoretical ratio is a symptom of bad port design. Although the use of excess air does improve mixing, as reported by Rummel, the advantage of this improved mixing is almost invariably lost on account of the increased waste-gas losses.

Since the goal is the design of a port which permits of efficient combustion without the use of excess air, further discussion is left until port design is considered.

Pressure in Working Chamber

The results of the experiments in which furnace pressure was the main variable gave very strong confirmation both of theory[3] and of the practical observations of Chesters and Thring. The value of pressure control and of furnace operation at high pressures is becoming increasingly widely known, but there is still a tendency to trust the visual appearance of flame at the doors rather than the indications of the pressure-recording instrument. Discrepancies between the appearance of flame at the doors and the readings of the instrument are generally due to interference with flame direction (either through splitting of the flame by high-velocity air or through deterioration of the gas port), and the present experiments have shown that such a false appearance of high-pressure conditions leads to reduction of air infiltration in the same way as a high pressure recorded on the instrument. There is therefore some justification for control by eye based on the appearance of flame at the doors, but such an amount of flame sting with a low roof-pressure reading implies misdirection of the flame in some way which is likely to cause abnormally short life of the linings. Also, visual indications are applied to some extent by flame quality. Strong indications have been obtained that the use of high roof pressures leads to a lower roof temperature, whilst maintaining the increase of heat transfer to the hearth (*see under sub-section* " Roof Temperature ").

Gas Calorific Value

Little further discussion on gas calorific value is required at this stage, since the results are qualitatively in accord with theory. Improved efficiency with high-calorific-value gas may be expected on account of higher flame temperature and diminished waste-gas volume.

Port Design

Port design affords the greatest field for discussion. It is now generally agreed that it is desirable that the air and gas should mix efficiently at as early a stage in their passage through the furnace as possible. Factors affecting mixing and combustion have been fully discussed by Rummel,[4] who points out that it may be assumed that combustion is effectively instantaneous once mixing of gas and air is complete, and that the degree of mixing is the criterion of general combustion efficiency. According to Rummel the most important factors favouring rapid and efficient mixing are :

(1) A large difference between the velocities of gas and air.

(2) High velocities of both the gas and the air streams.

(3) Excess air.

In the subsequent discussion these factors will be referred to as factors (1), (2), and (3). Now it is clear that these factors are to some extent mutually incompatible, for there cannot be a large difference between the velocities of gas and air if these velocities are both high. Again, excess air, however desirable from the point of view of securing rapid mixing, increases waste-heat losses to an extent which, in general, entirely offsets the gains brought about by rapid combustion, as is seen from the experiments now being reported. The only exception to this is where the port design is so bad that excess air is essential to secure even a moderate degree of mixing and combustion.

These opposing factors may account for many of the conflicting views held by practical operators. Also, the application of the above principles has to be modified by practical considerations, such as the necessity for the same ports to handle waste gases as well as ingoing gases, and the limitations imposed by available gas pressure.

Considering factor (1), maximum mixing takes place when a high-velocity jet of gas is ejected into a stationary volume of air (or *vice versa*). The mathematical treatment of such a system is complex, but the simplified treatment of Davis[6] can be made to show that the distance required for complete mixing of gas and the theoretical quantity of air for combustion is proportional to the diameter of the port from which the gas issues. This would also follow from dimensional analysis.* Consequently, to ensure rapid mixing according to factor (1), one port should be small and the other large. In practice the size of the small (*i.e.*, the gas) port is limited by the permissible pressure, but this limitation can be circumvented to some extent by the use of two or more small ports instead of one larger opening. In other words, the effective diameter of the port (governing mixing length) is reduced without reducing the area (which would require higher pressure). This suggests a return to the port design adopted in many early furnaces—some of which are still in operation—in which there were two or more gas ports. This type of construction lost favour

* This reasoning, expressed in a private communication (reference 7), neglects the effect of chemical reactions which take place (*e.g.*, combustion) and of any difference in viscosity, but it can be taken as a simple basis for further study.

because of constructional and maintenance difficulties, but with modern systems of water-cooling and improved refractories these objections may not be any longer valid.

The greatest amount of mixing by factor (2) will take place when two high-velocity streams impinge at an angle, but in practice the air velocity cannot be made high, since the air port must be fairly large in order to handle the volume of waste gases produced. The various systems of mechanically variable ports which can be changed at each reversal to accommodate ingoing fuel gases and outgoing waste gases more appropriately have not so far found favour, on account of mechanical and other difficulties.

The results from the experimental furnace must now be considered in the light of the foregoing remarks. Let us consider what we may expect when port design is systematically altered. Starting with a small gas port and large air port, we should anticipate that increase of the size of the gas port will bring about diminished efficiency, owing to the reduced mixing by factor (1). This does take place, as is shown by Figs. 9 and 10. On the other hand, the same effect is brought about by diminishing the size of the air port (the gas-velocity/air-velocity ratio is diminished). At the same time this reduction in air-port area does, by increasing the air velocity, increase mixing by factor (2), particularly if the ports are set at an angle. This effect does not counterbalance the reduced mixing by factor (1) until the size of the air port is substantially reduced. Thus, in Fig. 12 we see that efficiency is much diminished by decreasing the air-port area from 170 sq. in. to 110 sq. in., but substantially recovers when the air-port area has been further diminished from 110 sq. in. to 30 sq. in. The difference between the 25-sq. in. and the 30-sq. in. ports in Fig. 12*a* is not significant ; in any case, two different port types are involved.

This effect is enhanced when 55% excess air is used ; in fact, under these circumstances the smallest air port gives the best results (*see* Fig. 12, lower part). This is probably due to increased factor (2) mixing, *i.e.*, velocity, and not to the excess air *per se* (factor (3)), since it has already been established that excess air is undesirable (*see* Fig. 1). Perhaps, however, the abnormal pressure effect caused by the splitting of the flame—discussed under a previous heading—cannot be ignored, though at the present stage it is not possible to assess the contribution of each effect.

At the same time, the effect of factor (1) is not entirely lost, as with the small air ports and high-velocity air the larger gas port shows up better (*see* Fig. 11), and, with a larger gas port, the smallest air port is definitely best, as is shown in Fig. 13.

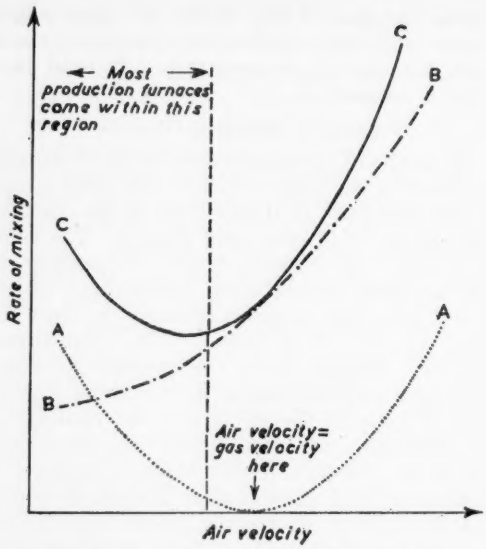


FIG. 14—Theoretical effect of air-port area on the mixing of gas and air

There appears to be an intermediate region of air-port areas which gives minimum efficiency. Some indication of this was given by a statistical investigation of the effect of gas-velocity/air-velocity ratio in which all the experimental results were used, but the result was by no means clear, except that there was a suggestion that this minimum occurred when the air velocity was somewhat less than the gas velocity. The view that such a minimum may exist is supported by a simple analysis of Rummel's theories as illustrated in Fig. 14, which shows the changes in rate of mixing, S, which may be expected with changes in air velocity, v_a, and any constant gas velocity, v_g. Curve A represents the effect of factor (1), in which the mixing effect is a minimum (zero as regards factor (1)) when $v_a = v_g$. Curve B represents factor (2), under the influence of which S increases with the square of v_a, since mixing by factor (2) is primarily dependent on kinetic energy. Curve C represents the total rate of mixing as influenced by v_a, and shows how the addition of the effects illustrated by curves A and B leads to a minimum at a value of v_a which is less than v_g. Further discussion of the theory behind Fig. 14 is given in an Appendix.

The conclusions on port design may therefore be summed up as follows :

(1) To ensure rapid mixing of gas and air the furnace should have either a small gas port and a very large air port, or a large gas port

and a small air port. Intermediate areas are not desirable.

(2) A furnace designed to use a relatively large gas port and small air port should have the latter directed towards the gas port. This is, of course, already done in nearly all furnaces, whatever the size of the ports.

(3) The air port should surround the gas port as far as possible.

(4) The small port (either gas port or air port) should be divided into two or more very small ports rather than have only one opening.

Practical Considerations

The evils of having to design the ports to take both ingoing fuel and outgoing waste gases are mentioned in almost every paper on open-hearth furnaces. Since the "one-way" furnace does not yet appear to be available, this fact must be faced, and it means that the second alternative suggested, namely, the large gas port and very small air port, may not be a practical proposition. A small air port demands the provision of air under pressure, and, whilst this presents no great difficulty, the exhaustion of waste gases through such small ports would demand a very high draught in the outgoing regenerators and flues, thus increasing infiltration to an undesirable extent. The only solution would be the use of variable ports; many such devices have been suggested, but so far none seems to have achieved permanent favour. Another method of securing the advantages of high air velocity is to admit a jet of air into the centre of the gas flame, as is done in the Loftus port,[8] or to have a high-velocity jet of air admitted through the "tongue" separating the normal gas and air ports, as has been tried by Cartlidge.[9] Such devices do not seem to have survived very long.

With the limitations imposed by existing practical conditions, the best port design would appear to be one which gives either of the following:

(*a*) A high-velocity gas jet with immediate access to a large volume of air, the horizontal velocity component of which is small; such a port would be a development of the present Maerz design, but with a small gas port and large air uptake.

(*b*) Moderate-velocity gas and high-velocity air, with the air almost surrounding the gas port and sharply inclined towards it; this port would be a development of the Venturi port, but with a more restricted air port, and would demand the provision of air under pressure.

Consideration might also be given to a revival of the system of two small gas ports, provided that water-cooling or other means of preserving equality between the ports was used.

The next series of investigations to be conducted on the experimental furnace will be a study of port developments along these lines.

It is interesting to compare the conclusions reached from the present series of experiments with those of Chesters and Thring, which were based on experiments on production furnaces. The experiments on the small-scale furnace support the views of Chesters and Thring regarding correct air/gas ratio and the desirability of relatively high pressure in the working chamber of the furnace. At first sight there would not seem to be the same agreement regarding port design, since Chesters and Thring obtained higher heat transfers to the bath with the *K* and *S* furnaces at Templeborough than with the *H* furnace, and reference to their Table I shows that the furnaces with large gas ports gave higher heat transfers. Study of the Templeborough data, however, shows that high heat inputs were also associated with the high gas flows permitted by the larger gas ports with the limited gas pressure available, so that, in the absence of the characteristic curves for these furnaces, the port designs cannot be directly compared on the basis of similar flow conditions. The conclusions on port design reached by Chesters and Thring are also affected by the ability to use the high heat inputs on production furnaces only if the port design gives vertical stratification, and thus a layer of relatively cool air between the flame and the roof. The fundamental views on port design reached from the experiments at Shelton do not conflict with these conclusions, which were reached from experiments with ports of different shape working under a range of experimental conditions limited by the necessity for maintaining production, rather than (as in the present series of experiments) on ports of different size working under a wide range of experimental conditions.

Heat Distribution in Hearth and Roof

In addition to the study of the effect of variables on the amount of heat transferred to the hearth, attempts were made to study their effects on roof temperature and in the distribution of heat transfer in the various zones of the hearth. The methods for roof-temperature measurement and for hearth-heat-transfer analysis have been outlined in Part I. Up to the present the amount of useful information which has been obtained from these measurements is not great, but it is

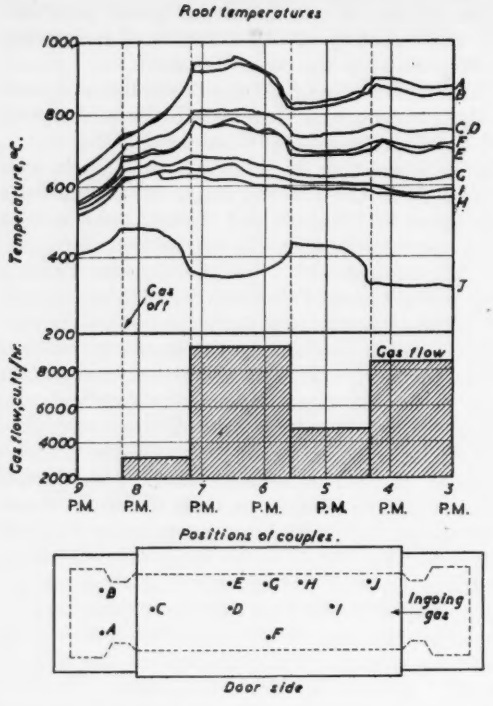


FIG. 15—Variation of roof temperature with gas flow

thought desirable to report the nature of the results so far obtained.

Roof Temperature—The full number of roof thermocouples used for these experiments was ten, but great difficulty was experienced in maintaining the couples, particularly at the hot end of the furnace. Towards the end of the campaign, when relatively high temperatures were being attained, it was decided to concentrate on the measurement of heat transfer to the hearth, and attempts to maintain ten thermocouples in commission were temporarily abandoned. During the present rebuild, opportunity is being taken to build-in much more robust thermocouples, and it is hoped that during the second campaign a complete picture of roof-temperature conditions will be obtained.

The original object of the measurement of roof temperature in detail was to determine the conditions which gave (*a*) the most even roof temperature and (*b*) the minimum increase of roof temperature whilst the heat transfer to the hearth was increased. This programme was vitiated by the fact that since the furnace is worked as a " one-way " furnace the roof at the ingoing end is consistently cooler than that at the outgoing end. The degree to which the roof temperature is uniform, therefore, greatly depends on the gas flow—at very low gas flows the temperature at the ingoing end rises and that at the outgoing end falls (*see* Fig. 15). Increases of gas flow giving greater heat transfer to the hearth increase the difference between the temperatures at the ingoing and outgoing ends, whilst also increasing the average roof temperature.

Figures 15 and 16 show the results of experiments carried out during the middle period of the furnace life, when all the thermocouples were operating. Figure 15 illustrates the effect of high and low gas flow on the distribution of temperature over the roof. In the absence of reversals the ingoing end (curves *H*, *I*, and *J*) is naturally cooler than the outgoing end (curves *A*, *B*, and *C*). Nevertheless, at high gas flows the difference is much greater ; in fact, reducing the gas flow—whilst causing a drop in temperature at the outgoing end—actually causes a rise in temperature

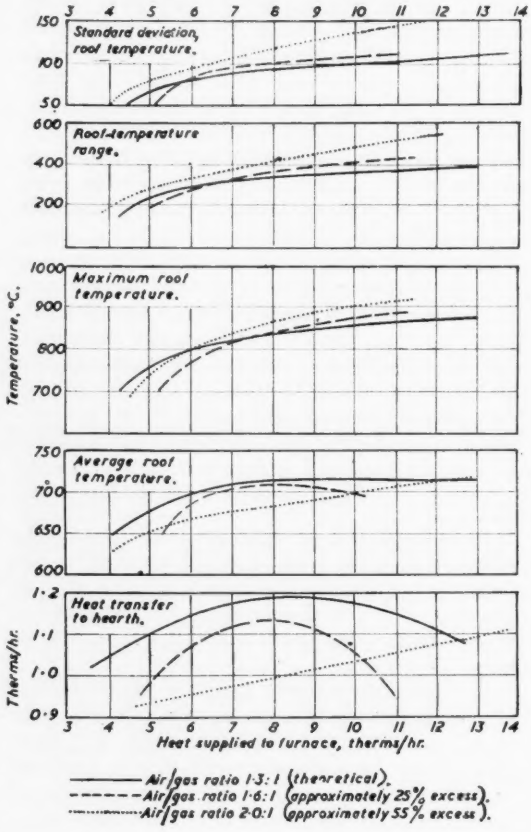
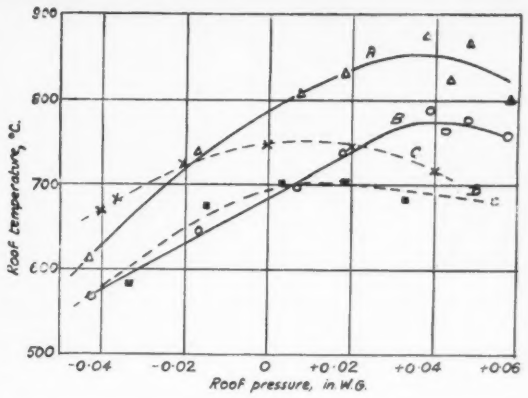


FIG. 16—Roof-temperature conditions, runs 89 to 125 (26-sq. in. port)

at position *J*, owing to the combustion zone moving nearer to the ingoing end, which not only brings a rise in temperature there but raises the flame locally. Figure 16 suggests that—apart from conditions of very low heat input, corresponding to impossible conditions in practice—unevenness of roof temperature increases with (i) heat input to the furnace and (ii) increase of air/gas ratio. However, it is not advisable to attach great significance to these figures, as they apply to only one set of experimental conditions, and these conditions were those giving a decrease of heat transfer to the hearth at the highest gas rates. When this decrease took place, roof temperature still tended to rise, and this supports the argument that excessive gas flow should at all times be avoided.

Figure 17 shows some results obtained with gas-port size as the chief variable, and these suggest that gas-port size has little influence on the distribution of heat over the roof.

On the whole, it was found that almost all conditions leading to higher heat transfers to the bath led also to higher roof temperatures. The only exception was in the case where roof pressure was made the variable, and the effect of changes of pressure on roof temperature (all other variables being kept constant) is shown in Fig. 18. It should be pointed out that the term " average " used in connection with this diagram is not a true guide to the average roof temperature, as the number of thermocouples from which this average figure was obtained varied according to the number of couples which were in working order. The thermocouples at the ingoing end survived longer, and therefore the average figure is weighted towards the lower temperature at this end. However, since the curve of the roof temperature at the outgoing end follows the average, these curves can probably be used as a qualitative guide to the behaviour of roof temperature at varying pressures.



Curve	Roof temperature	Gas-port Area, sq. in.	Air-port Area, sq. in.	Gas calorific Value, B.Th.U. /standard cu. ft.
A △	At outgoing end	9	170	180
B ○	Average	9	170	180
C ×	Average	26	170	150
D ■	Ingoing half of roof	26	30	180

FIG. 18—Effect of roof pressure on roof temperature

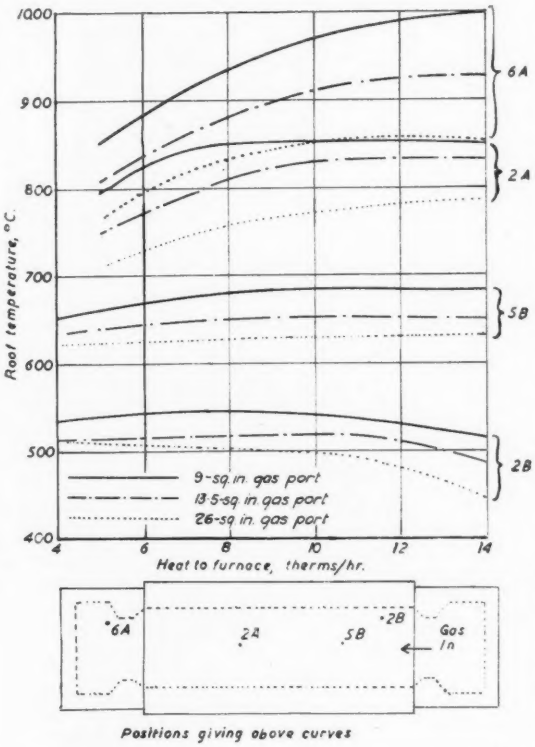
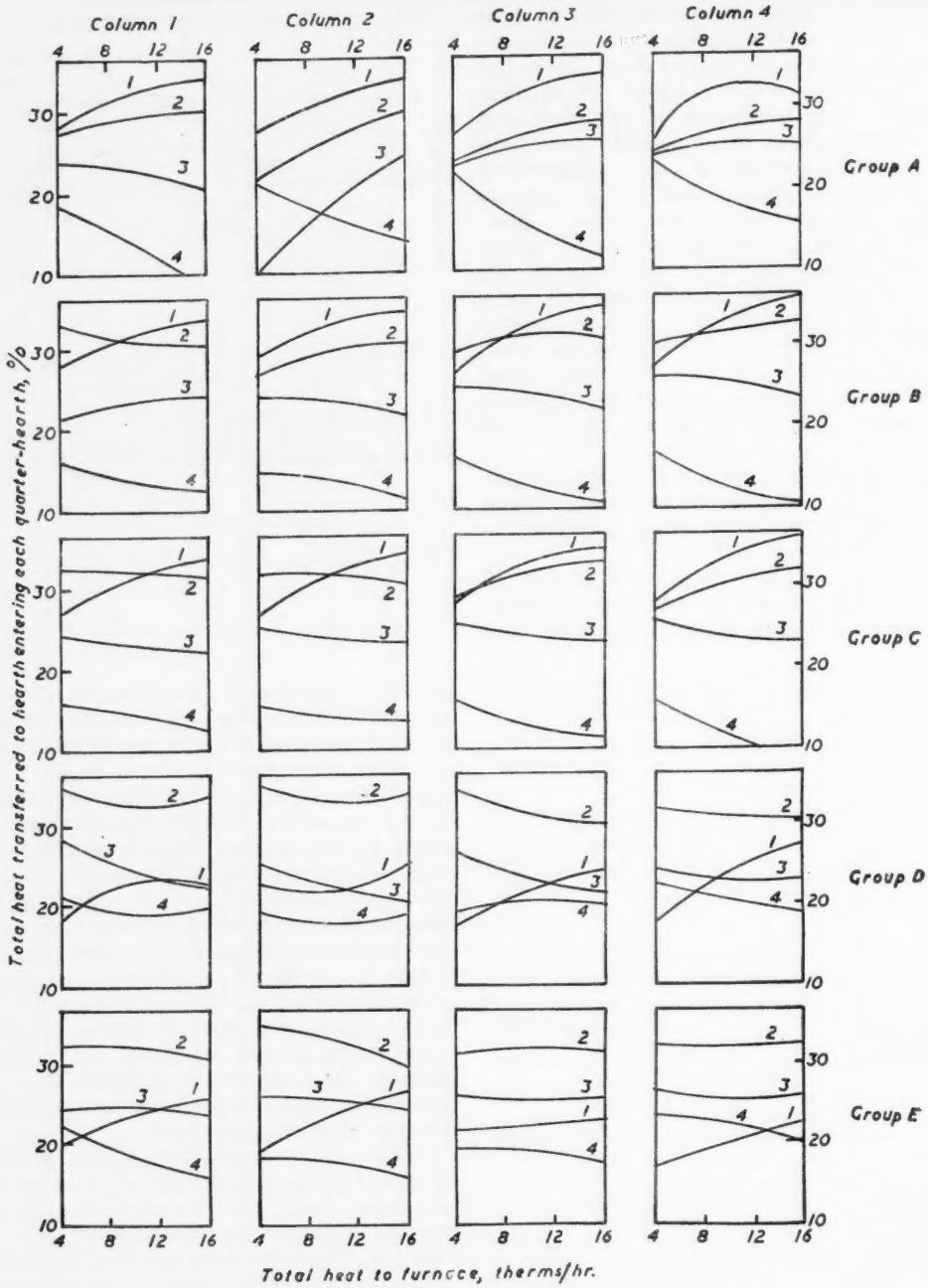


FIG. 17—Effect of gas-port size on roof temperature

These curves suggest that working the furnace at a high pressure under the roof is a method of obtaining high heat transfers to the hearth without excessive roof temperature (*cf.* Figs. 5 and 18). This conclusion can only be regarded as tentative until more consistent measurements of roof-temperature distribution are available.

The shape of the curves in Fig. 18 is at first sight remarkable, but the maxima can perhaps be explained as follows : At the lowest pressures a large amount of cold air is being drawn into the furnace, and this lowers both the heat transfer to the bath and the general furnace temperature,



	Group A	Group B	Group C	Group D	Group E
Gas-port area, sq. in.	26	13½	9	26	9
Air-port area, sq. in.	170	170	170	25	30

FIG. 19—Distribution of heat along the hearth with 15% deficiency of air (column 1), theoretical air (column 2), 25% excess air (column 3), and 55% excess air (column 4). Port designs Nos. 1 (groups *A*, *B*, *C*, and *E*) and 2 (group *D*). The curves designated 1 refer to the quarter of hearth length nearest the outgoing-gas end, and the curves designated 4 refer to the quarter of hearth length nearest the ingoing-gas end; curves 2 and 3 refer to the intermediate quarters

including the roof. As the pressure rises, the amount of cold air drawn in diminishes, causing a rise both in roof temperature and heat transfer to the bath. At zero pressure under the roof there is, however, sufficient cold air being drawn in under the doors to lift the flame. As the pressure increases further, the air infiltration is reduced to an extent which causes a further rise in the temperature of the furnace atmosphere which is sufficient to cause the flame to drop, thus bringing about a slight fall in roof temperature while the heat transfer to the bath still rises.

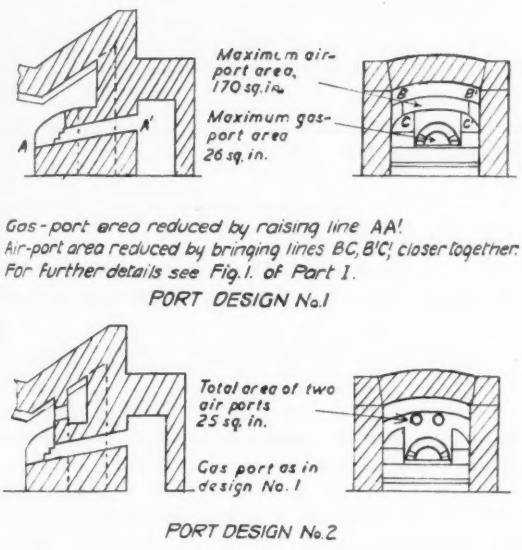

Maximum air-port area, 170 sq. in.
Maximum gas-port area 26 sq. in.
A
A'
B
B'
C
C
Gas-port area reduced by raising line AA'
Air-port area reduced by bringing lines BC, B'C' closer together
For further details see Fig. 1. of Part 1.
PORT DESIGN No. 1
Total area of two air ports 25 sq. in.
Gas port as in design No. 1
PORT DESIGN No. 2

FIG. 20—Principal port designs used in the experiments

Distribution of Heat along the Hearth—It was originally thought that the separate analysis of the amount of heat transferred to the four quarters of the hearth (measured along the length of the hearth) would be likely to give a very clear picture of the rate of combustion along the flame. In practice the difference in the heat transferred to each quarter is less than expected (in contrast to the difference in roof temperature along the furnace). This is no doubt due to the fact that the area of hearth at the incoming-gas end of the furnace (which is below a relatively cold portion of the roof) receives heat radiated from the flame and roof at the hot (outgoing) end of the furnace.

Nevertheless, some interesting features were observed and these are illustrated in Fig. 19, which shows the analysis of heat distribution in the hearth for several operating conditions. It will be seen that :

(1) Under most conditions the hearth at the outgoing (hot) end of the furnace receives more heat, as might be expected (groups *A* to *C*).

(2) The exception to observation (1) is where the air port is small ; the resulting high air velocity in such circumstances seems to have aided combustion to such an extent that the second quarter of the hearth (measured from the outgoing end) receives the most heat (groups *D* to *E*).

(3) With normal ports (groups *A* to *C*) the only circumstances in which the hearth nearest the outgoing end does not receive the most heat is when the ingoing gas flow is so low that there is no difficulty in making combustion complete towards the middle of the gas path. Even so, the slower combustion when the gas port is large (with a large air port) means that in group *A* the first quarter-hearth receives the most heat under all circumstances.

(4) Excess air leads to a more uniform distribution of heat under all circumstances, *i.e.*, earlier combustion is secured, but at the expense of overall net heat transfer.

The above observations are all in accord with the conclusions on port design discussed earlier in this paper, and with the views on mixing expressed by Rummel.*

The principal port designs used in the experiments, referred to throughout the paper as " port design No. 1 " and " port design No. 2," are illustrated in Fig. 20. Unless otherwise stated, port No. 1 was employed.

IV—ACKNOWLEDGMENTS

The authors desire to express their thanks to the Management of the Shelton Iron, Steel and Coal Co., Ltd., for the continued provision of facilities for the work ; to Mr. J. H. Cook for assistance with the experimental work during the later stages ; to Mr. M. W. Thring for helpful advice at intervals ; to Miss P. J. Arnold, Mr. R. L. Brown, and Mr. J. Pearcy for assistance with calculations and statistical investigations of the results ; and to the Electroflo Meters Co.,

* There are a few anomalies in the curves shown in Fig. 19, *e.g.*, the crossing of curves 3 and 4 in group *A*, column 2. It is difficult to explain an anomaly such as this except on an assumption that during this particular group of experiments there was some interference with the free and uniform flow of water through the hearth tanks.

Ltd., for the loan of pressure-controlling and measuring instruments.

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Appendix—*A Tentative Theoretical Derivation of the Curves in Fig. 14*

In the main text of Part II the possible existence of an air-port-area/gas-port-area ratio giving a minimum combustion efficiency is mentioned, and Fig. 14 is a simple graphical illustration (based on two factors governing mixing) showing how such a minimum could originate. A simplified mathematical background to Fig. 14 is that the two factors governing the rate of mixing, S, are (i) the difference between the air velocity, v_a, and the gas velocity, v_g, and (ii) the individual velocities v_a and v_g. This may be expressed as :

$$S = k_1(v_a - v_g)^{n_1} + k_2 v_g^{n_2} + k_3 v_a^{n_3} \quad \ldots\ldots\ldots(1)$$

In equation (1) Rummel's third factor, excess air, is not taken into account, as the experimental work has shown that excess air is so undesirable from the thermal point of view that any improvement in mixing owing to such air is not likely to result in higher efficiency.

Further detailed study of equation (1) is limited by the absence of more precise knowledge of the nature or numerical value of the constants and indices k and n. However, by making a series of simplifying assumptions it may be shown that this equation can be used as a basis for the construction of a diagram such as Fig. 14.

It may reasonably be assumed that the indices n_2 and n_3 in equation (1) are equal to 2, since energy is the main principle involved. It is more difficult to assess n_1, but the assumption that this index is also equal to 2 gives a curve in qualitative accord with observed facts, results in a real value of S whatever the relative values of v_a and v_g, and so simplifies the following reasoning that 2 is taken as the first assumption regarding the value of n_1.

Thus we have :

$$S = k_1(v_a - v_g)^2 + k_2 v_g^2 + k_3 v_a^2 \quad \ldots\ldots\ldots(2)$$

For any constant value of v_g,

$$\frac{dS}{dv_a} = 2k_1(v_a - v_g) + 2k_3 v_a,$$

and equating this to zero shows that the minimum value of S occurs when

$$v_a = \frac{k_1}{k_1 + k_3} v_g.$$

If as a further simplification the constants k are taken equal to unity, the minimum occurs when

$$v_a = \tfrac{1}{2} v_g.$$

Departures from the value of 2 taken for n_1,* or variations in the value of k, lead to changes in the shape of the curves shown in Fig. 14, particularly in the "production-furnace" region of low v_a ; for instance, a minor maximum as well as a minimum may appear in curve C, at a value of v_a approaching zero. This does not affect the argument that air-port areas only moderately larger than that of the gas port are likely to lead to combustion conditions of minimum efficiency.

It is probable that under all circumstances the rate of mixing under conditions of high v_a (the right-hand area of Fig. 14) is much higher than with the low values of v_a corresponding to practice in present-day furnaces, but to attain such high values of v_a in a melting furnace demands relatively high-pressure air and special ports for handling waste gas, *e.g.*, a non-reversing furnace.

The above argument does not take into account the effect of the relative angle of the ports. If the air and gas are directed towards each other, mixing will be improved, particularly at high velocities. It is clear that there is still a wide field for further investigation of the influence of port area on mixing, by both a mathematical study of the theories involved and by experimental work on models and small-scale plant.

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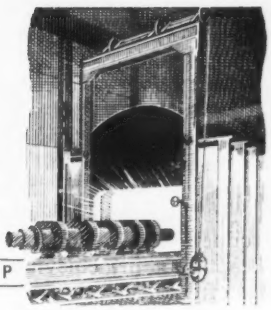
* For example, it might be argued that since the length of a turbulent mixing region is proportional to the diameter of the port (*see* reference 6) and v is proportional to d^2, the value of n_1 should be $\frac{1}{2}$.

IRON AND STEELWORKS ENGINEERING

THE IRON AND STEEL ENGINEERS GROUP



THE IRON AND STEEL ENGINEERS GROUP

REPORT OF THE SECOND MEETING

THE SECOND MEETING OF THE IRON AND STEEL ENGINEERS GROUP of the Iron and Steel Institute was held at the Offices of the Institute, 4, Grosvenor Gardens, London, S.W.1, on Wednesday, 11th December, 1946. Mr. W. F. CARTWRIGHT (Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd.), Chairman of the Group, presided, and Dr. C. H. DESCH, F.R.S., President of the Iron and Steel Institute, was also present.

AT THE MORNING SESSION a paper was read by Mr. H. J. KNIGHT on "Lubrication in Iron and Steel Works Engineering." The discussion was opened by Mr. L. D. COLAM, who compared British and American practice in Lubrication.

THE PROCEEDINGS OF THE AFTERNOON SESSION will be published in the April issue of the *Journal*.

PROCEEDINGS OF THE MORNING SESSION: 10.30 A.M. to 12.45 P.M.:

The **Chairman:** This is the last occasion on which papers for the Engineers Group will be read without being preprinted. I do not think that a paper on lubrication has ever previously been read before the Institute. There has been a big change in almost all the big steel companies as regards their attitude towards lubrication; in the old days it was a matter of a big kettle with some black oil and grease blocks, but now many of the large steel companies actually employ a lubrication engineer.

Mr. H. J. Knight, who is to present the paper on this subject, has just returned from a visit to the United States, where he has seen the latest practice, and, as he knows our own practice as well, his paper should be very interesting indeed.

Mr. H. J. Knight (Shell-Mex and B.P., Ltd.) then read his paper.

LUBRICATION IN IRON AND STEEL WORKS ENGINEERING

*By H. J. Knight**

THE reason for my visit to America was to investigate and examine modern methods of rolling steel strip, wide beam, etc. This, at first view, would appear to be far removed from association with petroleum products. Therefore I should explain that, through experience in this country, we are fully aware that for high-speed production, petroleum products—in the form of lubricating oils and greases—form an extremely important factor in this connection.

Many people present will have had experience of American rolling mills in this country, in spite of the fact that there are only a few, and it might be assumed that experience on such mills in England should be sufficient to cater for any new mills projected for Great Britain. However, in common with all engineering practice, new discoveries and new methods are constantly being evolved, and the period of warfare through which we have recently passed has undoubtedly

* Shell-Mex and B.P., Ltd.

accelerated advance in design and methods of production.

During my first few days in American steelworks it was quite apparent that the main thought in the minds of American steel producers is speed of production, consistent with quality. I couple these two items in fairness to our American friends as, although always striving after speed of production, they do maintain quality. If we accept that in achieving speed of production no loss of quality ensues, then this feature is the one which we should consider the most important. From the petroleum-product point of view it is extremely important, as inevitably the mechanical conditions for really high-speed production are different from those obtaining in respect of slow-speed production.

With this in mind it is imperative that production should be continuous, and this means that the plant should be capable of operating for long periods under arduous conditions. I repeat *arduous conditions* as, although theoretically the plant should be designed to cater for the conditions expected in a steelworks, it seems inevitable that circumstances arise—especially in the rolling of steel—which have not been foreseen, with the result that overloading and extremely adverse conditions often prevail.

It will be appreciated that, on paper, a steelworks lubricant should cover all items of plant, and I think I am correct in saying that in a works of this type practically all forms of mechanical equipment are employed, ranging from power generation, all forms of gears, pumping machinery, turbo-blowers, crushers, process plant, heat-treatment plant, etc., any one item of which could form the subject of a separate paper. Therefore I shall endeavour to deal with only some features of steelmaking, and even then not very intimately with any given item.

Blast-Furnaces

I shall commence with blast-furnaces. It is well known to most people here that there are a number of extremely important points of lubrication on a blast-furnace, and in the past lubrication has been effected mainly by hand. This generally means that efficient lubrication is entirely at the discretion of one person, and, he being human, when climatic conditions are bad it is possible for this work to be done, to put it mildly, indifferently. In addition there is always present the hazard to life, in that many parts are difficult of access and noxious fumes are present. Therefore, to achieve speed of production—or perhaps I should say continuity of production—it is essential that the necessity for repairs should be reduced to a minimum. We all know that very few bearings fail because of an unsuitable lubricant, but many fail for the lack of a lubricant. Therefore it is logical, if one is to achieve continuity of production, to ensure that the likelihood of failure owing to lack of lubricant is eliminated. With these points in view, in America most steelworks are standardizing on a centralized grease system for the mechanical lubrication of all parts of a blast-furnace. This might at first seem to be an expensive piece of equipment to instal, but if this cost is placed against that of maintenance I am assured by those concerned in America that the cost is infinitesimal and is more than worthwhile.

Obviously it is quite a specialized job to arrange for all the various parts to be lubricated from a centralized system. Further, it might well be argued that if the system breaks down all parts are without lubricant, whereas under the old method maybe only one or two would suffer. Since, however, it very often happens that if one or two parts fail this means a shut-down, there seems to be little argument in favour of the old method against the comparatively new one. The Americans have studied this problem very closely, and I have seen batteries of blast-furnaces working with extreme efficiency when equipped with centralized grease systems. The grease system and the pipework and general fittings must be of extremely robust design. A minimum, measured quantity of grease should be applied at regular intervals and the selection of the grease must be intelligently determined. Since my object was to learn something of the method of application and type of product used, I shall give the leading features which are desirable when considering the lubricant, having already briefly described the reasons for the particular type of application.

Consistency of Grease

Since of necessity the length of pipeline is considerable, it is desirable to have a grease of soft consistency at normal temperatures and which is easily pumpable at the low temperatures likely to be experienced on an exposed site. On this subject I would mention that many of the steelworks in America experience far lower climatic temperatures than those usually met with in this country, and they have found it expedient to encase the main supply lines to the top of the furnace with a steam tracer line. It should be noted that the steam line should not be in direct contact with the grease pipeline but

adjacent to it, the whole being encased in some form of lagging, which will ensure that the temperature will be sufficient to keep the grease at a consistency which will prevent undue build-up of pressure. By this means, of course, an adequate supply of grease is delivered to the working parts. It may or may not be necessary to use this artificial heating in Great Britain, but for the cost involved and in view of the vagaries of our climate I consider that it would be advantageous to instal the equipment complete with this heating device, which could then be used if found necessary. Actually this is purely a safety measure, as we can manufacture grease in this country which would work efficiently at the lowest temperatures likely to be experienced here.

Since the grease should be easily pumpable over a wide range of temperatures, the film strength of the grease cannot be disregarded. Therefore this feature should be controlled by the manufacturer at the highest film-strength figure possible. In this direction it is considered that a lead-base grease will give all the desired characteristics.

It will be seen from the statements made that speed of production, low maintenance costs, and absence of hazard to life can be catered for by the installation of a centralized grease system for this important item of plant. Some of the American blast-furnaces are fitted with fully automatic centralized systems and others with hand-operated centralized systems. The former is, of course, more expensive, but since one unit can be coupled-up to feed a number of furnaces the cost is divided.

Steelworks

Gas Regulators

Passing from the blast-furnaces through to the melting shop, and for the time being disregarding the grades of lubricants which are necessary for such plant as cranes, chargers, etc., I feel I should mention the gas regulators, which are usually—in fact practically always—hydraulically controlled. In many cases in America ordinary light hydraulic oils have been used, but these have frequently been found to cause trouble. The trouble can generally be attributed to oxidation, giving rise in turn to rapid increase in viscosity, thus causing sluggish control. On examination it was found that this, in the main, was due to contamination of the oil with furnace gases containing a percentage of hydrogen sulphide. The hydrogen sulphide combines with the unsaturated constituents of the oil and accelerates sludging. To alleviate this trouble it was found that a highly refined solvent extracted oil was necessary. By this I mean that the oil contains virtually no unsaturateds; thus there was no combination with the hydrogen sulphide. In addition, it has been proved by experience that it is advantageous to take the breather from the storage tank outside or away from the contaminated atmosphere.

Although a good solvent extracted oil is successful, this description is hardly adequate, as viscosity and other characteristics are of great importance and manufacturers of regulator equipment are quite well versed in the needs for their particular plant. I wish to make the point that manufacturers and oil suppliers treat this item very seriously indeed, and in America special thought is being given to oils containing oxidation inhibitors and other additives.

Electrical Equipment

We now come to a very important part of steelworks lubrication, namely, that of the electrical driving unit of the rolling mills. This item is usually left entirely in the hands of the electrical engineers, and, as continuous production is impossible without lubrication efficiency in this direction, I feel bound to comment on the subject. I confine my remarks to the lubrication of electrical driving units of the rolling mills, including blooming, slabbing, and continuous-mill motors.

When electrical drives were first utilized, both in America and in this country, the bearings were lubricated by the simple loose-ring oil system, and this worked quite well—and for that matter is still working well—where speeds were normal and the loading not too extreme. As, however, the aim today is for continuous production, it is vitally essential for this power unit to be able to work for long periods without trouble, there being very little time allowed for ordinary running repairs and maintenance. The lubrication must as far as possible be positive and not dependent upon the human element.

The electrical manufacturers have realized this, and the modern method is to have a separate oil-circulating system supplying oil to the motor bearings. This is not a new idea and has been a natural development, but as the older types of motors run on a very small quantity of oil in the sump of each bearing the tendency was to utilize a very small oil-circulating system. In America they have found that it pays to instal a generous oil system with adequate continuous-cleaning and filtering equipment. The question of deciding the quantity has been to some extent arbitrary, and practice has been the guide. Taking the average, I find that if the estimated number of

gallons of oil supplied to the bearings per minute is multiplied by 30, the resulting figure will be found to be quite adequate.

It may be asked why the figure 30 is used. This has been arrived at by testing the settling rate of an oil contaminated with small quantities of moisture and adventitious material, such as scale dust, etc. The motor bearings and electrical equipment are usually isolated quite effectively from the actual mill and there is not a great deal of contamination, but inevitably enough to warrant serious thought. The figure 30 really refers to a precipitation rate of 30 min., meaning that in that time the usual contamination experienced in this class of equipment can be settled and filtered out efficiently and with actually a little time to spare. Precipitation rates of up to 60 have been used, but these have been found to be unnecessarily luxurious.

In the selection of the particular oil the electrical manufacturers, both in America and in this country, from long experience have found that this should be governed by the following factors :

(1) Low viscosity, to avoid fluid friction and to assist in the rapid dissipation of any frictional heat.

(2) Resistance to oxidation, since the oil is expected to remain in service for a very long period.

(3) Good demulsibility. Everybody knows that an oil which is hygroscopic will form an emulsion and accelerate oxidation and the formation of sludge, which can restrict the flow to the bearings and cause innumerable difficulties.

As regards factor (2), a great deal of research has been carried out by the petroleum industry to combat oxidation, especially in connection with steam-turbine lubrication. Therefore, although these are not exactly comparable, it has been proved that a correctly refined oxidation-inhibited oil will give a far longer life and a greater factor of safety than an orthodox mineral oil.

The initial cost of the oil for this important part of the equipment should not be regarded as the only guiding factor, since the total lubrication cost of the electrical equipment, based on the tonnage produced, should be infinitesimal.

Gears and Pinions

It is the rule in modern American rolling mills to supply the lubricant to gears and pinions from a central system, and the lubricant is called upon to lubricate gears, pinions, and bearings. In many cases one system and one oil is supplied for all the gears, pinions, and bearings (bearings only of gears and pinions) from one system. Obviously, for instance on a five-stand tandem mill plus a slabbing mill, the pitch-line velocity of the gears varies immensely, and theoretically a different oil should be used for the higher-speed gears (such as those of the final stand on a tandem mill) than for those on a slabbing mill. In addition the bearings would require an oil of lower viscosity than that on many of the gears. Since to isolate the bearings would bring in constructional difficulties, the lubricant supplier is faced with the problem of supplying a grade of oil which is adequate for gear lubrication as well as for the lubrication of the bearings. The lubricant for the gears should be capable of withstanding shock loads and often extreme pressure, whereas the bearings require a fairly light lubricant which would have a reasonably high film strength in order to be capable of dissipating frictional heat as rapidly as possible. Thus, theoretically, a lighter oil is required for the bearings than that for the gears. In the past a compromise has been effected by supplying an oil which, although slightly on the light side for the gears, is rather on the heavy side for the bearings.

As regards the application of the lubricant, the majority of American users arrange for this to be pressure-fed to the meshing point of the gears. This is quite an important point, as I think it is now well established that for gears running with a pitch-line velocity of 900–1000 ft./min. it is possible to effect fairly efficient lubrication by means of gravity or even splash application, but for speeds in excess of 1000 ft./min. pressure feed is essential.

Even when the lubricant is fed by gravity to the top of the pinions and gears and the oil is collected by distributing trays and led closely into the meshing point, associated forces alone can prevent an adequate supply of oil to the working part of the gears. To extend the point, I would say that the oil flowing from the distributing trays would, to a great degree, be thrown by centrifugal force to the underside of the tray. In addition, windage could mitigate against an adequate supply of lubricant to the surfaces which so urgently need its assistance.

To revert to the statement that in the past a compromise has been effected in the selection of a lubricant, in America today—and for that matter in certain instances in Britain—on modern plants an advance has been made in that, by the use of selected additives, oils can be used which are correct in viscosity for the bearings and at the same time have a film strength, or resistance to shock loads and extreme pressures, which will

cater for the arduous conditions obtaining on gears and pinions. I think I am correct in stating that for the majority of gears and pinions on modern rolling mills in America this additive type of oil is utilized. I found that the most popular type of oil was one containing lead soap. These leaded oils, although often referred to as extreme-pressure lubricants, have not the resistance to extreme pressure which oils compounded with active compounds possess. Nevertheless, the leaded oils have a greater resistance to extreme pressure at shock loads than an orthodox highly refined straight mineral oil.

It may be asked why, since the conditions are recognized as being extremely arduous, lubricants which have abnormal resistance to extreme pressure are not employed. The answer to this is that inevitably the oil in the system is subjected to contamination by adventitious material, such as scale dust and water, and most—if not all—of the active extreme pressure additives which are used for other purposes would combine with the water, etc., and either form an emulsion or undergo a chemical change, which would be deleterious to at least the bearings. Therefore, at the moment, it is wise to exclude these active additives. I should mention at this point that a correctly blended leaded oil sheds water quite as well as straight mineral oils, which are usually noted for this feature.

Viscosity and Type of Oil

In connection with the viscosity and type of oil, the leading points to be reviewed in gears and pinions are (i) the type of gear, *i.e.*, single or double helical, spur, or worm, (ii) the maximum and minimum pitch-line velocities, (iii) the material, or tensile, strength of the gears, (iv) the ambient temperatures, (v) the method of application of the oil, and (vi) the type of bearings, *i.e.*, sleeve or anti-friction. It must be assumed that the loading in all cases is taken care of by the manufacturers of the gears and that all the usual factors of safety have been observed.

Quantity of Oil Required

Following a decision as to the type and viscosity of the oil to be used, the next feature to be considered is quantity. In America this figure is to a great extent determined by past practice. It is also compared with a calculative figure, and the average figures used for the determination of the quantity of oil required for a typical continuous mill are calculated as follows :

Gears—Assuming a viscosity of 1700 sec. (Redwood I at 100° F.), allow 0·25 imperial gal. per minute per inch of gear width for either up-stream or down-stream application.

I repeat my previous statement that direct pressure-feed of oil into the meshing point of the gears is almost standard practice in America for steelworks gear systems.

Bearings—For sleeve-type bearings allow 0·5 imperial gal. per minute per square foot of projected area.

For anti-friction bearings allow 0·46 imperial gal. per square foot of projected area, *i.e.*, area of shaft in the bearing.

Added together these figures will give a figure expressed in gallons per minute. This figure is then multiplied by the estimated precipitation rate, and for this particular installation the multiplication factor would lie between 35 and 40.

The precipitation rate is arrived at after consideration of the conditions prevailing, *e.g.*, if it is anticipated that large quantities of water and possibly scale may find their way into the system a greater precipitation time should be allowed, but on the average a figure of between 30 and 40 should be adequate to allow for precipitation and filtration. Of course, this figure can be influenced by the type of filtration units being used, but the above figures are given for a typical five-stand tandem rolling mill.

The usual operating temperature of the oil is 100° F., that is, the oil is maintained at this temperature throughout.

There are many other features which have to be considered when designing a centralized oil system, but most of the usual rules known to engineers apply, and there is hardly time on this occasion to go into details.

The oil system supplying lubricant to the gears, pinions, and bearings also quite often supplies oil to the screw-down mechanism, and in some cases to the screws themselves. Obviously, if the main object is continuity and high-speed production, a great deal depends upon this oil system and it is imperative that it should be carefully designed and that very adequate filtration arrangements are incorporated. In this connection it should be borne in mind that it should be possible to remove the heating elements or coils for periodic cleaning and repair. This is a feature which has sometimes been overlooked and which by some people is considered unnecessary. It is all the more important where space is limited. Where there is ample space, in America a duplicate storage and filtering system is installed so that the complete system of oil can be changed from one to the other, the one which is empty being cleaned and, if necessary, repaired without any interruption to continuous production. This may

seem rather luxurious, but it appears to be the trend in America, and I think the extra initial cost is soon repaid.

Back-Up Roll Neck Lubrication

The majority of large continuous mills in America are fitted with anti-friction bearings to the work rolls, but in many cases the back-up bearings are of the oil-film type. As regards the latter, quite a number of mills are arranged so that one oil system supplies oil to all back-up-roll bearings. This is quite apart from the gear-lubrication system.

The oil in general use in America is similar in characteristics to that used for similar purposes in Great Britain. One very essential factor for this class of work is that the oil must have good demulsibility and must be resistant to oxidation. With only one or, possibly, two exceptions this oil is a straight mineral oil and does not contain any extreme-pressure additives. Nevertheless, in view of higher speeds and possibly higher loading, thought is being given to the use of additive types of oils, such as leaded oils. As stated previously, most of the mills in America have one system supplying the same oil to all stands, but the most recently installed mills have two and sometimes three systems to cater for the increasing speeds encountered on the intermediate and final stands. This, we must all agree, is correct procedure, but it involves a greater initial outlay. If, however, speeds and continuous production are the main consideration, it is really essential.

With regard to the systems dispensing oil to the back-up-roll bearings, it is recognized in America that generosity in capacity is essential. The minimum precipitation rate allowed is 30 min., but more often than not this is exceeded, and, where space is available, this is a good feature ; in fact many engineers would like to see a figure of 40 min. as a minimum.

The main reason for the wish for a longer precipitation rate for back-up-roll oil systems is that the roll-cooling water seems inevitably to find its way into the oil system ; this can obviously cause a great deal of trouble, in spite of the use of oils with good demulsibility features. In America this point has been given a great deal of thought, with the result that the most modern installations have a complete spare oil system, so that whilst one system is in operation the other can be dehydrated and cleaned.

In the past, on high-speed continuous merchant mills, where the oil-film bearing is extensively used, the same oil system supplied oil to the roll bearings as well as to the gears. However, with the increased speeds now being achieved, the trend is towards separate oil-systems, that is, one for the mill bearings and one for the gears. This is a distinct advantage, as it is thus possible to do justice to both items.

I must point out here that several mills in America operate with anti-friction back-up-roll bearings which are lubricated by means of grease, but the oil-film bearing is the more generally utilized.

With reference to work-roll bearings, by far the greatest number are of the anti-friction type and are lubricated by grease. The grease is supplied from an automatically controlled system and is arranged to supply practically continuously or at predetermined intervals, depending upon the particular conditions. When this method of lubrication was first used in America a large pumping unit was installed and was expected to supply grease to the majority of the grease points on a complete mill. This naturally entailed very long pipelines and a build-up of high pressures. Experience showed that it was far more efficient to split up the grease systems and to have one unit supplying grease to, say, the slabbing mill and the first and second mills, another unit supplying the tandem mill, and a further system for such items as run-out tables, etc. This has proved to be economical, both from the point of view of consumption of grease and from the all-important angle of continuity of production. Some steelworks have gone one further and installed a spare pumping unit at a central point. In an emergency this can be cut into any of the lines whilst repairs are being carried out to any given pumping unit.

I think that everyone will agree that the splitting-up of the pumping units has much in its favour and, other than the slight additional cost, very little against it. As it is of such paramount importance to avoid stoppages, all American steelworks carry a large stock of spare distributing manifolds so that when anything occurs to prevent the correct flow of grease to any given point, rather than attempt to effect what might appear to be a minor repair and on investigation develop into a major one, the complete manifold—which after all is quite small—is put in. This entails only a small amount of pipe-fitting.

In the larger steelworks a separate repair shop is maintained for repairing and testing grease pumps, manifolds, distributing valves, etc. A further precaution taken by some works is to have always on hand a portable grease pump, which can be connected into the main grease line at any point should a pumping unit fail. These units are comparatively inexpensive, and if continuous

production is to be regarded as the major issue something of this type is a necessity.

The types of grease used for the lubrication of these anti-friction bearings vary, but in most cases a lead-soap grade is employed. This has a high load-carrying capacity and should be very waterproof and of a consistency which is easily pumpable at all the usual temperatures encountered. In addition, it must be non-corrosive.

On run-in and run-out tables, the rollers of which are either independently driven by an independent motor to each roll or are driven in sections by motors driving shafts with bevel gears, the bearings of the rollers are invariably lubricated from the central grease system, utilizing the same grease as that used on the anti-friction work-roll bearings. Where the other method is used the gears are, of course, lubricated with oil, either in an enclosed bath or by pressure from a central oil system. In many cases the bearings of the gears are lubricated by grease from a central grease system.

All these methods are familiar in steelworks in this country and call for very little comment.

Fabric Bearings

Fabric, or synthetic-resin, bearings are well known in this country and are working very efficiently with only water as a lubricant. In American many mills are using this type of bearing, lubricated by water and grease. This, I admit, rather surprised me, as I believe one of the outstanding points made for the use of fabric bearings is the saving effected in grease consumption. I gather that the initial reason for the use of grease was because of acute wear on the thrust collars. It was found that by using a phosphor-bronze collar, with the fabric as the main body of the bearing, high tonnages were achieved, but obviously the phosphor-bronze collar needed lubrication and therefore grease was introduced to the fabric as well as the collar.

At this juncture it might be of interest to refer to some work which has been done in this country in connection with fabric bearings. First of all it was generally agreed that for given applications a fabric bearing showed great saving, provided that the peripheral speed of the neck or shaft was over 150 ft./min. and that clean water was the only necessary lubricant. However, in operation the following two features emerged :

(1) In order to overcome static friction it was advantageous to inject a grease before shutting-down and previous to starting-up.

(2) Incipient corrosion was observed on the roll neck or shaft.

As regards (1), it was found that it was necessary to be selective when choosing a grease for this purpose, and where loading was high an extreme-pressure type of grease was found to be advantageous. At the same time the grease had to be soft in consistency and capable of imparting an extremely thin film.

To some extent the addition of grease before shutting-down and before starting-up prevented rusting, but in certain cases where it was not considered necessary to use a grease a soluble oil was injected into the water before shutting-down and before starting-up. This method was found to be quite good, but incipient corrosion still occurred over a period of time. Eventually it was found that by the injection of a water-displacing rust-preventive fluid into the bearing before and after, not only was incipient corrosion prevented, but bearing life was increased very considerably. It is worthy of comment that in the particular case under review the water from the mill was being exhausted into a dock where contamination was virtually forbidden.

Quite briefly, where fabric bearings are to be utilized the following points must be considered :

(1) Can a closed water system be employed ? If so, a low concentration of soluble oil and water can be used with great efficiency.

(2) If the water is exhausting to a river or dock where contamination is important, the injection of a de-watering rust-preventive fluid is an advantage.

(3) Where a high starting torque is experienced, the injection of a grease is necessary. The amount required is so small that the fear of contamination troubles is not present.

(4) Where starting torque is high and it is possible to use a soluble oil in a closed system, if a grease has to be employed as well this must be of the soluble type so that there will be no adverse effect on the soluble oil and water solution.

Rolling Oils

This paper, as previously stated, must perforce be quite general, but I cannot conclude without mentioning one very important feature which is applicable only to continuous-strip rolling. My remarks in this connection are confined to the cold-reduction of this material.

The tonnage of cold strip produced by the Americans is immense, by virtue of their vast automobile production and the sheet required for the canning industry. In this country, I understand that the highest cold-reduction speeds are

in the region of 2000 ft./min. In America they are rolling at 3500 ft./min. and are at the moment completing mills which, it is anticipated, will roll at 5000–6000 ft./min.

At the present time the procedure, briefly, is to take the hot rolled strip, pass it through scale-breakers, through continuous pickle lines, cold-water-wash it, hot-water-wash it, dry it with hot air, pass it through an oiling machine, coil it, and then convey it to the cold-reduction mill. For commercial or automobile-body sheet, that is, other than for tinplate, the oil frequently used is a light straight mineral oil. The cold-reduction mill is usually a five-stand tandem. The oil used at the end of the pickle line is mainly for rust prevention and is not intended to be a rolling oil. The actual rolling oil varies considerably from a soluble oil and water solution to a mixture of mineral oil, palm oil, soluble oil, etc., and water. Each individual works seems to have its own ideas, but they all operate successfully, although the finish of the sheet varies. I should explain here that the rolling oil for this class of work is usually combined with the roll coolant, in that it is fed from a common system on to the rolls and on to the sheet. In most cases the sheet is taken direct to annealing pots without any cleaning. This, of course, does not apply to tinplate.

To revert to the end of the pickle line, it will be recalled that the wet sheet is dried by means of hot air. It is my contention that by doing this an iron oxide film is imparted to the sheet, which is then covered with an ordinary straight mineral oil. It is, I think, well known that a straight mineral oil is not a true rust preventive; therefore, apart from the oxide film given to the sheet whilst drying, if the coils of strip are kept in storage for any length of time corrosion can occur underneath the light film of straight mineral oil, and this can be rolled-in during cold rolling. This can affect both the lubrication of the work and back-up rolls as well as impair the finish of the strip.

As regards the first undesirable feature, namely, inadequate lubrication of the rolls, it would be hard to secure conclusive evidence as to the amount of wear caused by the iron oxide, but I am certain that it will be agreed that if it could be eliminated it would be an advantage.

With reference to the finish of the strip, it may be argued that most of the iron oxide is eliminated during annealing, but personally I have some doubt and feel that the black dust frequently found on cold-reduced strip after annealing is to some extent caused by this. Without a doubt an incorrect rolling oil can also be responsible for trouble in this direction.

In dealing with the rolling oil itself, before deciding on the particular rolling oil to be used, a number of features must be considered. The main ones I would enumerate as follows:

(1) Speed of rolling.

(2) Percentage reduction of the strip through each stand.

(3) Analysis of the steel to be rolled.

(4) Method of application of rolling oil.

(5) Desired finish of strip.

(6) Cleaning—if any—of strip before annealing.

It would be impossible to deal thoroughly with this subject in the course of this meeting, but from the remarks I have made I feel that at a future date the subject could be the basis for a great deal of interesting discussion.

With sheet for tinplate, in America the usual procedure is the same as for automobile sheet as far as pickling is concerned, but quite often, instead of using a straight mineral oil as a rust preventive, either neat palm oil or a mixture of palm oil and neat soluble oil is utilized. Apparently this is not done in order to take any greater precautions against rusting, but to assist in the rolling operations in the first and second stands. The rolling solution utilized for tinplate is either straight palm oil combined with the roll coolant, or neat palm oil is applied to the sheet, plain water being used as the roll coolant. There are also variations of palm oil, water, lactic acid, etc., used in conjunction with the roll coolant, serving as a combined rolling oil and roll coolant.

Palm oil is, without a doubt, an excellent rolling oil, but it has many disadvantages, apart from the question of world shortage of this particular material, and most American manufacturers would gladly utilize a different product, which it is felt would eliminate the very dirty conditions which prevail wherever palm oil is used.

This question is an extremely important one at the moment, as with the ultra-high speeds at which the mills are proposing to run it is considered that the rolling medium can influence the success or otherwise of speed of production.

I am confident that this problem is not beyond the ingenuity of the petroleum industry to solve. Already a great of deal work has been done, and it is hoped that this work will be sufficiently far advanced to cater for any ultra-high speeds which may eventually be encountered in this country.

DISCUSSION

The **Chairman**: Mr. Knight was faced with a big difficulty. He had a great deal to say, and it was suggested to him that he should cover the whole field. He said that it would be easy to read a paper on every single one of the six or seven aspects of the subject which he has covered, and he is hoping that at a later date individual papers will be read on such subjects as centralized grease-lubrication systems, roll oil, and so on. The purpose of his paper is to stimulate the investigation of the big field which has to be covered.

Mr. **L. D. Colam** (Farvalube Ltd.): I should like to thank Mr. Knight for a very interesting paper. He has dealt very largely with American lubrication practice, and it may be thought by some that we in this country are far behind America. Well, we are and we are not. We all know that in everything there is a wrong way and a right way of going about things. I am not going to suggest that we are going about things in the wrong way, but to my mind there are two obstacles to progress in this country which do not exist in America and which make it very difficult for us to go about things in the right way.

The first of these obstacles is, to my mind, an insufficient realization that lubrication is a science and a specialized branch of engineering. In America, oil is a key industry, and key industries always breed a number of experts and specialists. America is an oil-producing country; the average American schoolboy knows his oil-fields and knows the methods of production and the different grades, and quite a number of oil terms are familiar to him by the time he leaves school and starts to study for some profession. With this unconscious oil-mindedness on the part of the youth of America it is not surprising that a substantial percentage of men go in for lubrication engineering and specialize in their early twenties, so that American steelworks and large engineering plants have no difficulty whatever in attaching these expert lubrication engineers to their staffs. These lubrication engineers are highly paid. They effect for their companies great savings, not only in power consumptions, labour, lubricants, and bearings but also, owing to the elimination of troubles caused by faulty lubrication, in the form of increased tonnages, which pay very rich dividends.

It is very debatable whether in the average British steelworks there is scope for such men on a full-time basis and, if there were such scope, whether the men could be found. In this country the word "lubrication" to a schoolboy simply conveys a picture of something very messy done with an oil-can and a piece of waste rag; it is not surprising, therefore, that when boys leave school they have no idea whatever of specializing in that branch of engineering, and the very few who do are so rapidly absorbed by the leading oil companies that it is difficult for industrial concerns to get hold of them.

The leading oil companies can offer these men a much greater field for research; they have facilities for sending them all round the world to study conditions, as is instanced by Mr. Knight's extensive tour of America, and altogether they can afford them a more romantic life than a steelworks can offer. How, then, is this obstacle to be overcome?

To my mind, the obvious solution is for you to call on the leading oil companies to supply the technical information that you first of all require. I have never been able to understand why this is not done to a greater extent. It is largely done by some of the bigger works, but some of the smaller works seem loth to call in these technical experts. I have heard it said: "That is all very well, but if I call in a man from the X.Y.Z. Oil Company he is merely interested, directly or indirectly, in selling me X.Y.Z. oil." I do not expect any of you to believe that these men are trained and maintained by their companies solely for your benefit, but their benefits and their interests are so closely allied with yours that I do not think that any reputable oil company would employ a man who would mislead anybody for the sake of getting an order. To do so would entirely defeat the object for which vast sums of money have been spent on research. So much, then, for the first obstacle, which I think can be overcome.

The second obstacle is the short-sighted policy adopted on expenditure. To some of you who have recently been juggling with millions in considering new steelworks and new mills, the expenditure of a few hundreds or a few thousands on lubrication equipment is a mere drop in the bucket, considering the safeguard which it provides for very expensive plant, but what about all the numerous plants up and down the country where any improvements in this direction have to come from the maintenance grant and not from capital? In very many of these cases the engineer knows that by improved lubrication he could save himself a good many headaches, but, when he puts a recommendation before his board, in many cases the first question that he is asked is "How much is this going to cost?" When

he replies, he is told : " We have run this plant for twenty years without this, and we think that we can run another twenty years without it."

Obviously the first question that the engineer should have been asked was not how much it was going to cost, but how much it was going to save, and some of the savings which I have seen are so staggering that I hesitate to mention them. There have been quite a number of cases where the power consumption has been almost halved, and I know of cases where an expenditure of £800 or so has been got back, not in years or in months, but in weeks—in six to eight weeks in some instances. I know of one instance where the improvements effected in lubrication by using the right lubricants and applying them in the right way have resulted in a saving of £450 a month from an expenditure of £1,300.

What we want in order to overcome this short-sighted policy on expenditure is a more general knowledge of what can be achieved. I have said that I think that in some respects we are behind America. On the other hand, we are not behind in other respects : We have the lubricants here, we have the technical experts of the oil companies, and we have the lubrication systems available. We have to find some method of bringing what can be achieved by the combination of these three things more to the front. I should like to ask Mr. Knight if whether during his tour of America he came across any figures of the savings which have been effected by plant-lubrication engineers using modern methods ?

Mr. **H. J. Knight :** The point raised by Mr. Colam regarding the employment of lubrication engineers is very important. In every steelworks that I visited in America a chief lubrication engineer was employed, and in the larger ones assistant lubrication engineers were employed at the various mills. I raised the question of whether that was really worth while financially and discussed it with chief engineers and maintenance engineers, and with one accord they said that they could not do without them. Since they have so mechanized their methods of lubrication there, it is a whole-time job for a lubrication engineer to supervise them. They collaborate, of course, with their oil and grease suppliers, and they have a separate repair centre in every works for the maintenance of the lubrication equipment. I expect that many people in this country have seen quite good mechanical lubricators really abused after six or nine months, and it seems to be nobody's job to see that they are repaired and really working efficiently.

I incline to the view, therefore, that a lubrication engineer could be employed with advantage in a steelworks, provided that the steelworks was big enough to afford him sufficient scope. There is need, as Mr. Colam says, for greater collaboration between the engineers at a steelworks—with, if possible, the employment of a whole-time lubrication engineer—and the oil and grease suppliers, so that intelligent discussion between them can lead to the most efficient results.

Mr. Colam also raised the question of the savings which can be made. I have many notes of instances where colossal savings have been made. I have not had time to get them all out, but I happen to have one or two with me. In one case at a very large steelworks the mills and the pinion bearings were changed from hand lubrication to mechanical grease lubrication, and the saving, according to the figures given to me by that concern, was $104,904 (about £20,000) a year. That saving was assessed on the actual renewal of bearings throughout the mill and the pinion bearings, so that it is a pure saving on bearings of £20,000 a year. It sounds fantastically high, but it is the figure which they gave me, and on checking it as far as possible it seemed to be correct.

In addition to that rather large sum, saved purely by the installation of mechanical grease equipment, I was surprised to see the estimates of the power saving on that particular unit. Mr. Colam has mentioned that figures of up to 50% have been obtained. In this case the power saving was over 50%. It seems an extraordinary figure, but I saw the power charts, and I could not find any fault with the figures.

The **Chairman :** There are two questions which I should like to put to Mr. Knight. First, are lead-base lubricants in any way dangerous to those who have to handle them ? Are they dangerous to the people who use them or fill the oil into the sumps ? Secondly, he said that for a gear-box one had to choose an oil which was neither the best for the gears nor the best for the bearings, but was a compromise. Is not it possible in a modern gear-box to use roller bearings and run them on a separate system entirely from the gears ? Is it the view of the lubrication engineer that roller bearings should be run on grease, or, if you could stop it leaking out, could you run on oil ?

Mr. **H. J. Knight :** Technically, it would be very nice to use a separate oil for the bearings of a gear-box, but it would increase the manufacturing costs, and unless you separated the bearings

entirely from the gears you would get contamination from the oil from the bearings into the gears. In practice it has been found efficient to have the bearings incorporated with the gear-box, and the same oil can be used for the anti-friction bearings as for the plain sleeve bearings. I think that most of the gear units supplied with modern rolling mills are equipped with anti-friction bearings incorporated in the design of the gear-box, and the same lubricant is used for the gears and the bearings.

In the past, as mentioned in my paper, a compromise had to be effected, and an oil was used which was really too thick for the bearings and on the light side for the gears ; however, with the advent of additive types of oil, it is possible now to do justice to both and to get the maximum efficiency from the lubricant as well as from the gear and bearing parts of the unit, so that I think the answer to that question is that with the advent of up-to-date additive types of oil it is no longer necessary to make the compromise which had to be made in the past.

On the question of whether oil or grease should be used for roller bearings, roller bearings work just as efficiently with oils as with grease, especially in a unit where there is free exit for the oil to go back to the sump. It is efficient to use oil so long as one point is kept in mind, and that is that any oil chosen must be strictly non-corrosive in the presence of these rollers. As long as that is watched there is nothing to prevent the use of an oil for those bearings, although grease is also used in many cases. If grease is to be used on the bearings of a gear-box, it is necessary to make absolutely certain that very little grease gets into the gear oil, as obviously the viscosity of the oil would rise considerably in such a case.

Lead-base oils have been supplied and used very extensively in this country and to an even greater extent in America, but there has never been a single question from operatives in steelworks of any danger from them. I will go further and say that in making these leaded oils at our installations we have never had a single complaint regarding oil dermatitis or lead poisoning. We thought at first that it might be a difficulty, but it has never been a problem.

The **President of the Institute** (Dr. C. H. Desch, F.R.S.) : The Chairman has asked me to say a few words of welcome to this gathering. This is the second meeting of the Engineers Group. The first was very successful, and I am glad to see so much interest taken in the subject of this meeting.

The only section of Mr. Knight's paper of which I have had much experience is the last one, that of the oils used in rolling strip. The Rolling-Mill Research Committee has had work on this lately, and the experimental work on the determination of the coefficients of friction with different lubricants gave surprising results. I know that the oil companies have done an immense amount of research on the nature of the processes of lubrication, following the pioneer work of Hardy a good many years ago, but when lubricants were arranged in the order of the coefficient of friction the order was surprising, and it was very difficult to find any theoretical reason. The determinations were made very accurately by Dr. Orowan at Cambridge.

The lowest coefficient of all was given by talc, which is out of the question in practical use, but the order of the other materials was also unexpected. I imagine that since the time of Hardy a great deal of work must have been done by the oil people on the characteristics that are wanted in a lubricant. This is rather a special case, because rolling oil does differ from the lubricants used in bearings, but I should like to know whether they are able to put forward at present a definite idea as to the kind of structure that is required in an oil to make it an efficient rolling lubricant.

Mr. **H. J. Knight** : The main question raised by Dr. Desch is whether any details are available to determine the correct structure of an oil which is going to be used as a rolling oil, namely, for the reduction of cold strip.

Dr. **C. H. Desch** : Yes, why one oil is good and another is not.

Mr. **H. J. Knight** : There is no conclusive evidence at the moment, but we think that it is a matter of surface tension to a great extent, in that when considering the actual function of reduction of metals it is imperative to have water or some means of cooling the rolls. I think it is understood that in a rolling mill, where you are reducing the metal at high speeds, if you were to get a water phase between the roll and the actual interfacial friction point, water being a poor lubricant you could get metal-to-metal contact and pick-up of the actual strip on the rolls, which would give a bad strip and cause acute wear to your very expensive rolls.

It is quite easy to see, therefore, that what is important is to prevent that water rupturing the actual oil film at the point where the interfacial friction takes place. To do that a study was made of surface tension and the "wettability" of various oils and liquids, and it was found that by making a rolling oil which had peculiar surface-tension characteristics it was possible to keep the water back, or rather to maintain an oil film

at the point of contact rather than a broken film of oil and water. At the moment, therefore, it would seem that the leading feature when designing a rolling oil would be to keep the water back beyond that point where the friction is taking place. There is still, however, a great deal of work being done, and I think that in a matter of months, or possibly weeks, some conclusive evidence may be available to enable a paper to be written giving the real characteristics necessary for the success of a given type of lubricant as against another.

Mr. **W. Craig** (Davy and United Engineering Co., Ltd.): While I hesitate to add to or in any way criticize this excellent paper, I am very happy to be given the opportunity to pay my tribute to the author of a paper from which we all have profited. Mr. Knight has just returned from a visit to the U.S.A., where he would no doubt meet many clever engineers, and he has brought together in this paper the information he has obtained and given us the facts and the fruits of a long experience in the lubrication of machinery.

The problem of lubrication now occupies a very high priority from the point of view of the designer. Instead of being tacked on at the end of the job, as it so often was in the past, it now occupies a position of primary importance; in conceiving a machine one of the first things to think about is how it can be efficiently lubricated. That this condition has been brought about is due largely to the efforts of Mr. Knight and his colleagues, who have gone up and down the country spreading their propaganda and impressing on designers and users alike the importance of efficient lubrication. In this respect he and they have rendered excellent service to the iron and steel industry.

Mr. Colam has referred to the lubrication engineer. I have heard of such people in America, but I have never yet had their duties clearly defined to me, and perhaps we can have some information on this subject from Mr. Knight. In my opinion the success or otherwise of a modern lubrication system starts on the drawing board; that is where the mechanism is built up, and on the results of the designer's skill and experience the success of the whole scheme depends.

The choice between grease and oil presents itself very early. The use of grease goes back to the old fat blocks and the boy with the bucket and spade who put a little grease on the bearings and much more on the floor. The grease did its work on the bearing, but after use and having nowhere else to go it finally deposited itself on the floor and, when mixed with mill scale, resulted in the shocking mess so often associated with our steelplants. In later times we have become more clever, and we cover up all gears and pipe the lubrication points to a central station, and the grease is then applied by means of one of those efficient little pumping units, operated either by hand or by mechanical means, and in this way grease is becoming a little more efficiently used.

The use of oil, however, seems to present a much cleaner and more economical method of lubrication. Where the oil can be applied from a central source and, after use in the bearings, be returned again to the central source, filtered, cooled if necessary, and used over and over again, it would appear to be a more economical method of lubrication. This happy state of affairs, however, cannot be brought about without a great deal of care, skill, and experience on the part of the designer. He must see that all his casings are oil-tight and that his covers and bearings are properly designed, as otherwise the whole system can become a nightmare. Oil seems to ooze out of all sorts of unexpected places, *e.g.*, porous castings, bad welds, covers, bolt holes, etc. I seem to be very familiar with all the places where it can come out, but, given proper care and attention in the early stages of design, I am confident that a thoroughly oil-tight job can be made and that such a system is very efficient.

I consider that this paper will be a source of valuable information to all designers of steelworks machinery.

The **Chairman**: We should all be interested if Mr. Knight would say a few words on the duties of a lubrication engineer in the U.S.A.

Mr. **H. J. Knight**: The duties vary considerably with the size of the works. In one of the largest plants in America the chief lubrication engineer works very closely with the chief engineer's department and also with the maintenance department. In other words, he has two masters. If there is a question of the installation of some new plant, he is the man who will sort out the methods and even go into the costs of the equipment—which, I may say, comes as a capital charge—and therefore he has to serve the chief engineer in advising him on the equipment to buy in order to secure the lowest cost consistent with efficiency. When it comes to the question of efficiency, as he has another master, namely, the maintenance department, he dare not go simply for what is cheapest, or it will be reflected in high maintenance costs. To begin with, therefore,

he is really a go-between between the engineering department and the maintenance department, and it is his duty to examine and be familiar with all types of equipment for the application and handling of lubricants and for filtration.

In addition, he is responsible for the storage of the oils and greases. That may seem a very small point, but when the amount of lubricants used in a large steelworks is considered it will be realized that the storage of them is a very responsible job, as many thousands of pounds are spent annually on those products.

Moreover, he has to make a report over varying periods on the efficiency of the particular units which have been installed, and endeavour to make them more efficient by cutting down, or experimenting in cutting down, the consumption of lubricants dispersed through the mechanical grease lubricators and oil systems. Obviously he has to be familiar with the characteristics of the products he is using, and he gets most of that experience by collaboration with the oil and grease suppliers. I was very surprised and pleased to see—and I hope that it will happen in this country—that the lubrication engineers of the steelworks make frequent visits to oil refineries and blending establishments and discuss with the supplier even the actual blending of certain additives which they may think would be advantageous for a particular application. There is perfect collaboration between the supplier and the lubrication engineer. Of course, it is the duty of the lubrication engineer to advise the engineering department of any new developments in petroleum and to ascertain whether they can be utilized with advantage in steelworks.

An over-riding job of the lubrication engineer is to produce monthly figures showing the cost of lubrication per ton of steel rolled or per ton of steel made. Those figures are charted and closely watched. It is the direct responsibility of the lubrication engineer to see that the cost per ton is reduced to a minimum, consistent with efficiency.

I like the idea of judging the efficiency of lubricant not on the price paid per gallon or per ton but on the actual cost to the user expressed as the cost per ton of steel rolled or per ton of steel made. That in the past has been done in this country by the oil companies over here, and it is very surprising to find—and I have some figures on it—that frequently it is possible to use really expensive high-grade lubricants in a steelworks and at the end of the year find that the cost per ton of steel rolled has been reduced. The figures indicate that when the proper oil or grease is used, then, in spite of a possibly relatively high initial price, if the lubricant is correctly applied it will show a saving at the end of the year. In reply to Mr. Cartwright, therefore, I can say that the job of the lubrication engineer is to bring the cost of lubrication per ton of steel rolled down to an efficient minimum.

There is one other duty of the lubrication engineer, and that is that when a new mill or new equipment of any kind is being ordered the designs are discussed with the manufacturer before the order is placed. The engineer is then able to collaborate with the manufacturer in making the unit capable of being lubricated efficiently right from the beginning, instead of—as often happened in the past and as may happen in this country—a piece of equipment being purchased and then the lubrication engineer, or in this country very often the lubricant supplier, being asked to produce a miracle out of a barrel by providing something that will work. The collaboration of the lubrication engineer with the manufacturer of new equipment is a very important matter.

Mr. **E. T. Judge** (Messrs. Dorman, Long & Co., Ltd.): This subject has been chosen at a most opportune time and is certainly one of high importance; having listened to Mr. Knight's paper and his replies, I feel that we have made an exceptionally happy choice of a speaker to deal with it.

Mr. Colam drew a distinction between the two sides of the problem, namely, (i) the application of efficient lubrication to new equipment which is to be installed and (ii) the improvement of lubrication practice on existing plant. I think that I agree with him as far as (i) is concerned, and no doubt the point concerned will be taken care of, but as far as the improvement of lubrication practice on existing plant is concerned, we have to look at it from the point of view that to be able to spend millions on new plant for the steel industry we shall have to maintain and improve as far as we can the earning power of our existing plants. In my opinion, we should be able to improve the earning power of such plants to a very large extent by improvements in our general lubrication practice.

I do not think that there are any companies in the steel trade to-day who, faced with expenditure for the improvement of lubrication practice, are not going to ask what this expenditure is going to save. I think that in each case they will be satisfied by the figure of saving which they are given.

Mr. Knight described the centralized grease lubrication of blast-furnaces. Personally, I think

that in the past, particularly on blast-furnace tops, we have neglected to provide adequate access to parts to be lubricated by the old hand methods, and I hope that we are not going to ignore proper access being provided in the future if we go on to centralized systems; a blast-furnace top, as Mr. Knight points out, is not a very pleasant place to be in, and in addition to the general lubrication problem there is need for a periodic inspection and check-over.

The second question upon which I should like to ask Mr. Knight for enlightenment is the use of phosphor-bronze thrust-rings on fabric bearings. Can he tell me on what type of mills the Americans have found it necessary to go to phosphor-bronze thrust-rings? His suggestions as to the correct systems of lubrication to use on fabric bearings under various conditions are going to be of great value in improving efficiency on existing plants.

Mr. **H. J. Knight**: With regard to the type of mill where phosphor-bronze collars are used in conjunction with fabric bearings, it so happens that the recent instance which I saw was on a wide-section mill, and it applied to the horizontal roll bearings and not on the vertical roll bearings. I was interested in that, because some rather unusual claims (as I regarded them) were made as to the life of fabric bearings, but when I discovered that the thrust collar was virtually separate from the body of the bearing, and it was agreed that many thrust collars were worn out during the life of the main body of the fabric bearing, I realized that the figures given were not comparable with those for the life of a fabric bearing using only water as a lubricant.

That brings up the other point, of lubrication by grease. I made the statement in the paper that apparently as the collar had to be lubricated the body was lubricated as well. I can only assume that, because there seems to be no real reason for the supply of grease to the main fabric bearing, if we go on the experience which has been gained in this country, where in the main, other than starting-up and stopping, water is used as the lubricant throughout.

Mr. **E. T. Judge**: I think that some of the practical people here would be interested if Mr. Knight would explain how on a centralized system for blast-furnace lubricatcn the man operating the system is assured that grease has actually reached a particular bearing on the blast-furnace top, which may have to be flexibly connected to the mains system, with consequent risk of fracture of the feed pipe.

Mr. **H. J. Knight**: That is a very good point, one that I raised myself when viewing these centralized systems. The arrangement generally made there is that the bearing furthest from the system is coupled up electrically, and when the man is pumping by hand he goes on pumping until an electrical contact is made, which tells him that his maximum pressure has arrived at the furthest point from his central grease pump. When he knows that he has his required pressure at that furthest point, he can assume that all the bearings up to that point have received their charge of grease. Later I can give you some sketches of the actual apparatus, if these would be of interest.

The **Chairman**: Mr. Stock has a centrally lubricated blast-furnace complete with steam tracer line and can tell you some of the troubles with it.

Mr. **A. Stock** (Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd.): We have this centralized lubrication system and we have electrical control, so that when the grease reaches the top bearing a red light comes on in the hoist-engine house, where the pump is situated.

We have had a certain amount of trouble in connection with the tracer line. There does not appear to be any satisfactory means of telling exactly when the tracer line should be turned on. I do not know whether in America they have overcome this difficulty with any automatic-control device, but I imagine it would be a good idea to have an automatic-control device worked on a temperature basis, so that the tracer line could be turned on automatically when the temperature dropped.

Mr. **H. J. Knight**: I did not see any automatic device such as you suggest, but it sounds a very good idea. In America the question of when the artificial heat should be put on is merely judged by the person in charge by whether winter has arrived or it is still summer, so that it is left to the person in control rather than to any arbitrary temperature control. I agree that it would be a very good idea to have automatic temperature control, especially in this country, where it may be very cold one day and quite humid the next. I think that the people who manufacture these systems might give the point some thought, because the arrangement would not be expensive.

Dr. **F. T. Barwell** (National Physical Laboratory): On the subject of rolling oils, I was surprised that Mr. Knight put surface tension

first when considering the desirable properties for such oils. I think that consideration of what occurs during the rolling process will indicate that the coefficient of friction is the vital property. This is probably the reason why, as he says, palm oils are so extensively used as rolling oils in the United States. I appreciate the importance of keeping water away from the actual surfaces, but I consider this to be incidental to the achievement of a really low coefficient of friction. I emphasize the point because I think that Mr. Knight will agree that it is important from the point of view of attaining a high speed of rolling and at the same time getting a big extension per pass.

I should like to ask whether the thrust-rings which Mr. Knight mentions in connection with fabric bearings in the United States are just plain rings or whether they are grooved or divided into segments in any way.

Mr. **H. J. Knight**: The ring is a solid ring, and not made up in sections; it is a machined ring, not in parts.

In reply to the first question, in doing some experimental work both in laboratories and in practice, when we first tried to evaluate the properties required for a rolling oil we naturally assumed—as I think that most people have assumed in the past—that it was a matter of resistance to extreme pressure, as the pressures exerted there, virtually on a line contact, are very high; therefore the earliest work done was to try to make a product which had great resistance to extreme pressure, and thus assist lubrication. We were all very surprised to find that extreme-pressure additives did not facilitate rolling, but we discovered that it was a question of wettability—that is, getting the strip wet with the rolling oil—rather than any real effect of extreme pressure. It was a question of maintaining that film at the point where the interfacial friction occurs. As long as you can maintain the film there for that very small time it is possible to roll without any undue trouble, but the addition of additives to give greater lubrication features as far as pressure is concerned certainly did not give any promise of success.

It all came back to the nature of the liquid, and that is why I think that palm oil is to a great extent successful, because it is acidic and inclined to attack the metal and it gives wettability; actually, palm oil, and many of the oils which have been tried and found to be successful, have no great extreme-pressure properties, so that it is a question of reconciling lubrication with surface tension and disregarding to a great extent the question of resistance to that extreme pressure which we know exists at that point.

Dr. **F. T. Barwell**: That was not quite the point to which I was referring. I was referring specifically to the importance of the coefficient of friction between roll and work. For example, what, in Mr. Knight's view, would be the effect of the inclusion of 5% of oleic acid in a mineral oil?

Mr. **H. J. Knight**: It would be better than the mineral oil without the 5% of oleic acid, because it would wet the metal far better than the straight mineral oil and it would give a better coefficient of friction. The coefficient of friction is very important, not so much at the actual point where the metal is being reduced but as affecting lubrication between the back-up roll and the work roll, and thus preventing as far as possible that acute wear on the surface of the back-up roll and work roll. The oil has to come from the actual strip, unless it is applied in a solution form over the whole.

Dr. **D. Clayton** (Ministry of Supply): Dr. Desch has raised the question of the fundamental ideas of boundary lubrication as put forward originally by Hardy; nothing that Mr. Knight has said contradicts Hardy's view of the polar molecules as being the effective lubricants. I do not remember Orowan's results, but in regard to talc there may be some confusion, because talc is a solid material and with it the metals do not come into as close contact as they do with boundary films. The rolling oil and the palm oil mentioned by Mr. Knight contain the soaps, esters, and fatty acids which constitute the main series of polar molecules. These materials will also be excellent emulsifiers in some cases; by a similar action the same type of material forms an adsorbed film on the metal and keeps the water away from it, preventing corrosion. In the case of the fatty acids—the oleic acid already mentioned is one—they will combine with the metal and give soaps on the surface, which soaps are exactly of the type of the adsorbed soap molecules but formed *in situ*. The same polar molecules will in many cases lower interfacial tension between oil and water. It will be seen, therefore, that everything fits in with the classical view.

Mr. Knight raised an interesting point when he referred to the difficulty of a compromise between the different requirements for bearings and for gears. The solution which he offered was quite a logical one, *viz.*, lowering the viscosity of the oil for the bearings so as to lower the power consumption, and, because that would make the

lubrication of the gears unsatisfactory, putting in a good boundary lubricant, namely, the lead soap. That is the oil man's solution, but as this is an engineers' group I suggest that consideration might also be given by engineers to altering the bearings, adjusting the width or the sheared area, so that an oil of correct viscosity for the gears does not cause excessive heating or power loss in the bearings. It is better not to depend on boundary lubrication, even when this is improved by the use of lead soap, because whenever boundary lubrication occurs it is likely that wear also occurs, and it is not therefore altogether satisfactory during continuous operation. If the engineer and the oil man get together a satisfactory answer should easily be possible.

Mr. **E. Hallas** (The Brightside Foundry and Engineering Company, Ltd.) : Mr. Knight referred to pressure lubrication of the double helical teeth of the pinion-housing gears when pitch-line speeds of 900–1000 ft./min. are met with. I should have thought that with adequate air volume in the pinion-housing casing and an adequate volume of oil delivered to the trays now embodied in the modern gravity-feed system, the lubrication to the teeth would have been efficient with this system without resorting to pressure lubrication.

The mill designer has in the past worked very closely with the lubrication engineer in designing his plant. Close consultation with him at the present time is very necessary, as higher rolling speeds and heavier pressures on gear and roll necks are called for on mill plants of all types. Mr. Knight has referred to the lubrication engineer being called in at the end to provide a lubricating system on a plant which has already been made. As far as rolling mills are concerned I think that this has generally been a question of the capital cost of the equipment. In the past it has on some occasions been the mechanical lubrication system which has been deleted to bring down the cost.

The design engineer and manufacturer of the plant is very interested in the question of efficient lubrication, and, knowing what the plant will be called upon to produce, he does co-operate with the lubrication engineer in putting forward a very sound system of lubrication.

We have in this country a large number of small and medium-sized bar and merchant mills, where frequent roll or roll-stand changing is met with. Mechanical lubrication to the roll necks has been a difficult question in the past, as very often roll changing has been left to the mill crew instead of to mechanics.

If the time comes when we have a lubrication engineer on the staff of all steelworks and non-ferrous plants in this country, the efficient operation of lubricating systems will not rest with him alone. He will require the co-operation of all concerned, and there is scope now for a good deal of propaganda directed to the mill crews and operators generally in order to educate them in achieving the desired objective.

I should like to congratulate Mr. Knight on his presentation of a first-class paper.

Mr. **L. D. Colam** : I should like to enlarge on the reply given by Mr. Knight to Mr. Stock's question as to what indications are provided to judge when a steam tracer line should be used on a blast-furnace. There is a pressure-gauge at the source of pumping, which gives the operator a good deal of information. Apart from that he should be able to tell blindfolded, because if the weather conditions affect the pumpability of the grease he will obtain excessive pressure in a shorter time than is usual. If the pump usually requires, say, 20 pulls to build up the required pressure, and he finds that he builds up the pressure in five or six pulls, it is obviously because the pumpability of the grease has been affected by the cold weather. On the other hand, if the pump takes its 20 pulls and the pressure is not built up, the operator knows that he has a broken line somewhere and that the grease is escaping. He has these indications on the feel of the pump itself, apart from the fact that the pressure-gauge will give him the same information.

Mr. **J. Romney** (Messrs. W. B. Dick and Co., Ltd.) : We all appreciate that in view of the very wide field covered by the paper it has been impossible to go into very much detail in the time available, but there are details which would be very helpful, and, to give an example of the type of information which we should like to get from Mr. Knight, I shall state a hypothetical case. Suppose that I was running a rolling mill and I called in Mr. Knight and asked him for a rolling oil. He would ask me for all the particulars which he has detailed. He says that the choice of a lubricant will be affected by the speed of rolling, the reduction at each step, the analysis of the steel, the method of application of the lubricant, the finish, and the cleaning, if any. I would give him all that information, he would go away and presumably work out what is the most efficient lubricant for the job, and he would suggest a particular grade.

A week later I might ring him up and say that when I told him that the steel had a 0·2% of carbon I meant 2%, and, further, that I had not

intended to clean the strip but that I now intend to do so. What I want to know is, what influence would that new information have on Mr. Knight in his selection of the lubricant ? In what direction does the change in the analysis of the steel lead him to alter the lubricant ? In what direction would he alter the lubricant to cope with an increased speed of rolling ? It is probably too much to ask him to give us the details now, but perhaps he could include something on these points in his written reply to the discussion.*

We have heard a good deal about the terrific savings resulting from the introduction of improved lubrication methods. These seem to suggest that the original lubrication methods were very inefficient and that the lubrication engineer has been faced with a fairly easy problem ; he has found a bad method and has substituted a good one, and so has saved money. But what are we to do in those cases where the lubrication system has been improved and put right ? In what direction can we work ? Given an efficient lubrication system, we must look to improvements in the lubricants themselves. That presents some difficulty, unless we have a good deal of collaboration on the part of the operators and also the willingness to try new things. The oil man works out a new lubricant for a particular application. Usually it is not a shot in the dark ; he has a theoretical basis for it and it is probably supported by laboratory work and by practical work on a minor scale, but it is difficult to get the operator to give the new material a trial if the lubricant which he is using at the time seems to him to give satisfaction.

I suggest to operators that they should absorb rather more of the spirit of adventure. Obviously they cannot try everything that anybody suggests, but if they have reason to rely on the judgment of the oil man they should accept his suggestions and be willing to give a trial, even though the trial may ultimately turn out to be a failure, to some new product.

Dr. Clayton referred to the difficulty of having a central lubrication system for lubricating gears and pinions and also bearings when the different applications call for oils of different viscosities, which means selecting a light oil which will do for the bearings and coping with the necessity for dealing with the gears and pinions by introducing various additives. Dr. Clayton suggested the alteration of the bearings to fit the viscosity of the oil which would otherwise be suitable, without additives, for the gears. You get the same sort of thing in turbine lubrication, but in the other direction. Where you have a geared turbine, you use an oil heavy enough for the gears, though you could do better with a much lighter oil for the bearings.

Is it not possible that the centralized lubrication system is somewhat overdone and that it would be an advantage in these cases to have one lubrication system for the gears and pinions and an entirely independent one for the bearings ? I do not know whether this is mechanically desirable, but if it is it seems a better proposition than any attempt to alter the bearings in the way suggested.

Mr. **H. J. Knight :** I should like to deal with Mr. Romney's questions in the reverse order. He commented at the end of his remarks on the lubrication of a turbine or gear system and bearings, and said that it might be a good thing to separate them. My paper is concerned mainly with steelworks lubrication, and that is the tendency in America and over here—that is, to treat the gears and gear bearings (because they are integral together) as a separate problem rather than to try to include other bearings on that same system. That is the tendency, and, going still further, in certain rolling mills in the past it has been common practice to build a system which has to lubricate all the bearings and all the gears, but I would say—and I speak now of American practice—that almost every mill which is being built of that type has now two, and sometimes three, different systems. They isolate gears from general bearings, and that is technically correct and does not cost a great deal more.

I can appreciate the point Mr. Romney makes about altering the oil instead of altering the bearings. I imagine that the manufacturers of the equipment would have something to say about making the bearings fit the oil. I think that the oil industry has gone a long way along the road by experimenting with and finally achieving oils containing additives which fill-in that gap and thus avoid the need to use a compromise.

The questions on rolling oil are too long to answer here, and so I should like to accept the suggestion that the answers be given in the written reply.

A previous speaker asked about the lubrication of gears and/or pinions by pressure feed as against gravity. From experience in this country, confirmed by experience in America, I mention in the paper a pitch-line speed of 900–1000 ft./min. as being generally accepted by engineers as the point at which pressure feed should be applied and the gravity system discontinued.

* *See* p. 441.

As for the reasons for that, it has been observed that as soon as you get a high pitch-line velocity the tendency is for the oil to be thrown away, even though it is dripping or fed from the trays near the meshing point. Windage and centrifugal force will tend to exclude the oil rather than to take it in, and in some cases it has been observed that the oil is running down the distributing tray, and, instead of dripping straight off the bottom of that tray, it seeps underneath. The gear coming in an upward direction tends to throw that oil on to the underside of the tray rather than into the actual meshing point of the gears.

That applies only when the pitch-line velocity is over 1000 ft./min. The figure of 1000 ft./min. has been arrived at by observation in the field, and so I think that it is a good figure to take, as it will be only 100 ft. either way where this phase comes in of throwing the oil away from the meshing point of the gears rather than taking it to the part where the work is to be done.

Mr. **F. L. Smith** (Messrs. Dorman, Long & Co. Ltd.) : I was interested in Mr. Knight's description of the functions of the lubrication engineer, and it struck me that the lubrication engineer might be drawn from two classes, (*a*) an engineer with a chemical bias or (*b*) a chemist with an engineering bias. Whilst the chemist, with his more meticulous mind, might be very useful, an engineer would probably get on better with the maintenance engineers and the men. It would be interesting to know from Mr. Knight under what category come the lubrication engineers whom he met in the U.S.A.

Mr. **H. J. Knight** : That is an interesting point, and it is one in which I myself was interested. The general specification for a lubrication engineer is that he must be first of all a mechanical engineer with a knowledge of electrical engineering. They place mechanical first and electrical second. He must be young enough to be able to absorb a good deal of petroleum technology, and he must be able to direct the men under him, so that he can get the work done which has to be done.

Mr. **F. B. George** (Consett Iron Co., Ltd.) : I should like to add my tribute to the author for this very interesting paper. Although my life is not yet, I hope, in the sere-and-yellow-leaf stage, I well remember that in the mill in which I first worked we used suet for the rolls. It would make the mouth of the modern British housewife water to see the enormous quantities of suet which came into the mill each day. I do not think that it did the roll necks very much good, but the local rats had a good time, and when we changed from that practice, shortly after I went into the mill, one is so accustomed to receiving deputations in steelworks that I was surprised not to have the shop steward of the Amalgamated Union of Rodents coming to complain about the change of practice and to ask for some compensation !

However, we did change that practice, and we invited an engineer from one of the oil firms to come and give us some advice. He was very helpful to us, in fact, so helpful that I think he got into trouble with his principals, because he not only told us the correct grade of grease and what sort of grease blocks to use on our necks, but also told us that by reboiling the grease that we reclaimed and adding something known in the trade as " Yorkshire yellow fat " we could make satisfactory grease blocks ourselves. I think that he was told that his job was to sell grease to us, not to show us how to make it for ourselves !

After that we went on and tried all sorts of different systems of lubrication. We have, of course, proper oil circulation on pinions and pinion bearings, but we have tried all kinds of pressure grease systems for various applications on cranes and on hot and cold mangles and on roll necks, and our experience has been that the real difficulty is not to find the suitable lubricant but to find the foolproof system. It is a comparatively simple matter to design for a new layout a centralized system which will be satisfactory, particularly when dealing with oil, although even with a new layout there is sometimes a lack of forethought, particularly with regard to methods for indicating failure of the supply. I well remember an occasion on a modern continuous strip and bar mill which has a very efficient centralized oil-lubrication system ; the supply tank was fitted with a tell-tale to show how much oil was in and there was a pressure-gauge on the delivery line to the bearings, but there was no audible alarm. One of the return-oil pipes from the bearings, in a very inaccessible position under the housings, broke. Before anyone noticed that the oil was going down in the supply tank, about 1200 gal. of very expensive lubricating oil, which at that time (during the war) was almost unobtainable, went into the scale pit, and we were put to considerable difficulty and expense in recovering that oil again after it had been contaminated by water and scale. After that experience we fitted an audible electric alarm which showed at once any pressure drop in the system.

When you come to design a system for an existing plant, however, and particularly a grease system, you are met with great difficulty. I am thinking of such things as hot and cold mangles,

cranes, and chargers. Everybody in a steelworks knows the difficulty of adequate lubrication of cranes and chargers; they work continuously, and there is always a fight about them between the engineering and the production sides, the production side saying that there is no time to lubricate them and the engineering side saying that they are being ruined for lack of lubrication. The hot mangle, for example, is one of the most difficult things to keep properly lubricated.

There were two main difficulties which we came up against in grease systems. The first was to find some adequate method of showing that the grease was actually getting to the place for which it was intended, and the second was how to deal with, say, a hot mangle where there are large numbers of bronze bearings, many of which are not always in the best condition, and where the wear is different on different bearings. The difficulty there was to ensure that the grease got to the bearings that required it and not to those that did not require it or did not require it so much. There was a tendency for all the grease to go through to the bearing which was a little loose, and come out at the end. One system we tried was of foreign make, and during the war the difficulty of obtaining spare parts meant that in time it entirely ceased to function.

I think, however, that those difficulties can be overcome, and recently we have put a system on a plate mill which seems to be giving good results. This has been in operation for about six months. The bearings, and particularly the bottom roll-neck bearings, on that mill, which formerly were changed at the end of three weeks, now last about four months. Not only do we get the benefit from the bearings lasting longer and therefore costing less per ton, but also it is very much easier to keep the mill properly lined up and to roll a more satisfactory product. I do not think that anyone would think of putting down a modern high-production mill without properly designed centralized lubrication systems, but I still think that there is a great deal of work to be done in the designing of systems, particularly grease systems, for existing plant which will do their job.

Mr. **G. Foster** (Messrs. Dorman, Long & Co., Ltd.): I would like to support what Mr. George has said, particularly as it applies to the older type of section rolling mill.

Roll changing in such mills, even at the present time, may occupy from 8 to 14% of the mill's available time. The time required to fix and unfix grease pipes is additional to this lost time.

In small three-high mills the chocks are so close together that greasing by grease block is almost an impossibility. In such mills the introduction of fabric bearings, which renders lubrication, except by water, unnecessary, is welcomed by the operators.

Greasing of roll necks by grease blocks is not a satisfactory method. It is difficult to keep the grease in contact with the roll neck, however conscientiously the mill crew attend to their work.

On the motion of the Chairman, a vote of thanks was accorded to the author for his paper.

The meeting adjourned from 12.45 P.M. *to* 2.15 P.M.

Written Reply to Questions on Rolling Oils

Mr. **H. J. Knight**: As regards how the difference in the type of steel to be rolled would influence the selection of the rolling oil, I would say that in reducing a comparatively soft steel less work is done and that there is therefore a lower interfacial friction with which to contend. Where a tough steel has to be considered there is a greater amount of work to to be done and greater interfacial friction to be dissipated. This would obviously demand adjustment of the viscosity and the general characteristics of the rolling oil and/or the quantity applied.

As for the question of how cleaning of the strip or otherwise affects the selection of the rolling oil, the strip would require annealing after cold reduction and inevitably a certain amount of oil would remain on it. It is well known that some oils come away from the strip in the annealing pot, leaving no stain or carbonaceous deposit, whereas others will possibly leave a dirty surface, or even staining, which is difficult to remove. Many different types of oils can be used for the physical action of rolling, but comparatively few combine the virtues of being good rolling oils and are at the same time suitable for bright annealing.

Lastly, in answer to the question of in what direction does increased speed of rolling affect the rolling oil, from experience it has been found that speed of reduction directly affects the rolling oil, in that there is less time for the oil to dissipate the interfacial friction and the general operating temperature at that point is higher. Therefore, serious consideration has to be given to the viscosity, the amount of oil applied, and other characteristics of the basic oil and/or additives utilized.

Generally speaking, it will be realized that there are so many variables in connection with the cold rolling of strip that all features have to be considered as a whole when endeavouring to select a rolling oil.

80-in. Cold-Reduction Reversing Strip Mill at Messrs. John Summers and Sons, Ltd., Shotton, Chester

THE Davy-United rolling mill recently put into commission by Messrs. John Summers and Sons, Ltd., at Shotton, Chester, is the first important unit to be installed by this company since the war and forms part of their expansion programme. It is also a notable plant in that it is the largest four-high reversing mill for rolling steel strip in this country. Designed to cold-reduce hot-rolled pickled strip up to a maximum width of 74 in., it is capable of handling a maximum coil weight of 10 tons. A general view of the delivery side of the mill is given in Fig. 1, and the general layout is shown in Fig. 5.

The whole plant was designed and manufactured by Davy and United Engineering Company, Limited, and consists of the mill proper with tension reels on either side and with cone-type uncoiler, pinch rolls, and entry sticker on the entry side. The work rolls are 20½ in. in dia., and the back-up rolls 53 in. in dia., and 80 in. long on the barrels. The mill housings, each weighing about 75 tons, are of massive construction and are rigidly connected by heavy cast-steel separators at the top and bottom, and the base is bolted to heavy cast-steel shoe plates which are secured to reinforced-concrete foundations.

Questions of motor power and speeds of rolling were the subject of careful consideration, and finally a motor of 2,500 h.p., equated to rolling speeds of 500–1,000 ft./min., was decided upon. A direct-coupled 2,500-h.p. " Metrovick " motor, running at 93–186 r.p.m., drives the mill through spindles of forged alloy steel, with universal couplings on both ends with spring-balanced carriers. The pinion housing is of cast iron, of the totally enclosed type, and of very massive proportions. The mill pinions are of special steel, solid with their shafts, with double-helical machine-cut teeth, 20 in. pitch circle diameter, mounted on roller bearings, and continuously lubricated from a self-contained system comprising duplicate oil-pumps, one as standby with alarm in case of failure or breakdown.

The work rolls are of alloy forged steel, and are carried in Timken bearings; the back-up rolls are of alloy cast steel, rotating in Morgoil bearings.

Particular care has been taken with the design of the roll balance system to take up all clearances between the top working and the top back-up rolls and between the top chocks and the mill screws so that no lift occurs when the strip enters the mill. This is achieved by an arrangement of hydraulic cylinders and rams carried in the lower back-up chocks, the rams of these cylinders thrusting against the upper back-up-roll chocks. Pressure is also maintained between the top work roll and the top back-up roll by hydraulic cylinders mounted in the lower work-roll chocks, the rams of these cylinders thrusting against the upper work-roll chocks.

The work rolls and their chocks fit into vertical extension pieces of the back-up-roll chocks, and are held in position in this assembly by means of latches which fit into slots in the back-up-roll chocks. By slipping these latches out of the chocks the work rolls are speedily and easily removed and changed by a porter bar. Back-up rolls are changed by a C hook or, as an alternative method sometimes adopted with this type of mill, by the introduction of slide bars on which the roll chocks are mounted, enabling the whole set-up to be withdrawn from the mill by means of an electrically driven rig. The roll-change side of the mill is shown in Fig. 3.

Work-roll chocks and screws, chock slides, etc., are serviced from a "Farval" central pressure grease system. Back-up-roll bearings are lubricated from the Company's main oil-circulating system with alarms in case of failure or breakdown.

The coolant system for the work rolls consists of sets of spray nozzles on both sides of the mill which are connected up to an existing system, and so arranged through magnetic control valves that only one set of nozzles is operating at a time, depending on the direction of rolling.

The screwdown gear is operated by two 50-h.p. motors through double-spur and worm reduction gears with a ratio of about 1,000 : 1 and giving a screwing speed of 0·229 in./min. The motors can be used independently or as one unit by means of a magnetic clutch, thus permitting the adjustment of both or either screw during rolling operations. All gears are mounted on Timken bearings and

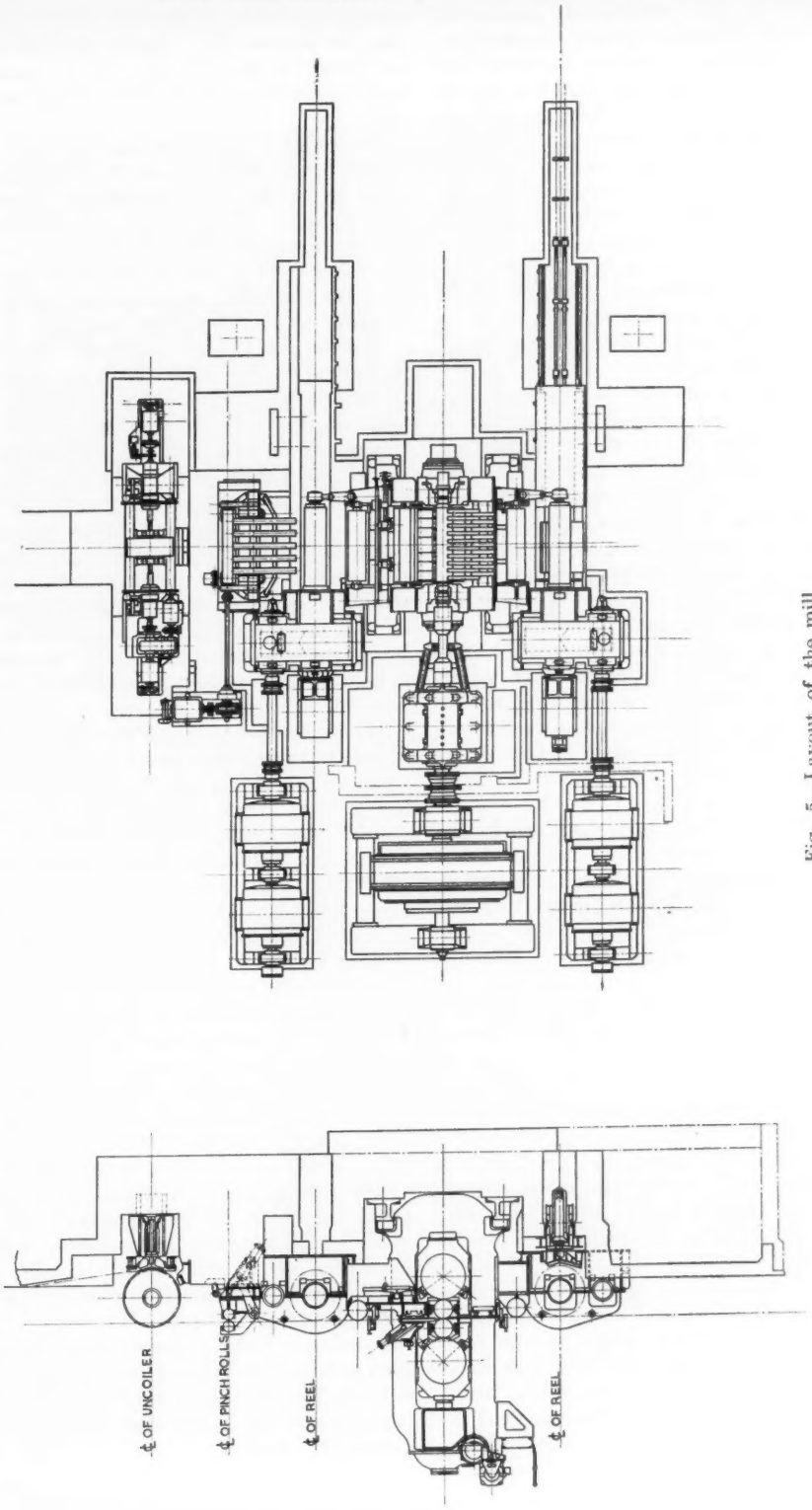


Fig. 5—Layout of the mill

are totally enclosed and continuously lubricated by means of a self-contained system mounted on the screwdown platform. A feature in connection with the screwdown mechanism is that the tops of the mill housings are extended to form the bottom part of the screwdown enclosures, and cast-steel top-half casings are bolted on, thereby ensuring that screwdown worm thrust is transmitted to the roll housing, and, incidentally, reducing possible points of oil leakage to a minimum.

A Selsyn transmitter is connected to the screwdown gear which gives readings to 0·001 in. on the control tables. Flying micrometers are fitted on either side of the mill, gauge readings being transmitted to the control tables.

The ingoing side of the mill is fitted with an entry-sticker guide, the purpose of which is to give the leading edge of the strip the forward movement which will enable the work rolls to bite and also to provide an ironing effect on the strip during the initial pass. It possesses an air-operated top platen with the bottom platen wedge adjusted to give the correct height for the pass line. Both platens are faced with strips of hard wood. Hand-operated side guides are also included to suit different widths of strip. Immediately before the sticker is an air-operated deflector plate which guides the leading edge of the strip on to the bottom platen and so between the work rolls of the mill itself. After the first pass the top platen is lifted clear of the strip. The entry side of the mill is shown in Fig. 2.

The whole of the auxiliary equipment previously mentioned and now described in greater detail forms a most important part of the installation and is designed to reduce to a minimum the manhandling of the strip throughout the whole operation. The result of this is a small crew for the mill, in this case only four men.

Each coil is rolled down a ramp on to a hydraulic lift which consists of a cast-steel cradle. Rollers incorporated in the cradle allow the coil to be rotated by the uncoiler cones, which enables the leading edge to be passed over the bottom pinch roll. The uncoiler is a pair of cast-steel cones, one of which is electrically driven and fitted with a magnetic clutch to allow the cone to be disconnected when the strip is in the pinch rolls. The cones are arranged to traverse so as to hold coils of different widths.

The leading edge of the strip is now passed to the pinch-roll machine. This is a pair of forged-steel rolls running in roller bearings, the rolls being 12 in. in dia. and 80 in. long on the hardened and polished barrels. The bottom roll is fixed and electrically driven, while the top roll is designed to swing upwards by means of air-operated cylinders so that the strip can be entered over the bottom roll. Hand-operated side guides are fitted in the mill side of the pinch rolls to keep the strip central while being fed into the mill.

Between the pinch rolls and the entry sticker is a light hinged table which can be elevated to give a fair lead to the strip and prevent it from sagging over the reel and deflector roll. This table is inclined at an angle, after the first pass, to allow the reel on the entry side to be used. Two deflector rolls are provided, one on each side of the mill, rotating on Timken roller bearings.

The reels on the entry and the outgoing side of the mill are of the collapsing type. Each is driven by two 500-h.p., 200–675-r.p.m., D.C. motors in tandem, through reduction gear.

The reel drum is built up of alloy cast-steel segments and is 24 in. in dia. and 80 in. long. Expansion and contraction are effected hydraulically.

The construction of the drum is such that it possesses the maximum strength to resist distortion during rolling operations. Besides this the reduction-gear side of the drum is supported back to the mill housing and a swivelling outboard bearing at the other end is also mounted on the housing face, giving it additional support to prevent deflection.

When the finished coil is stripped off the reel drum, it is supported by a hydraulic lift while the hydraulically operated stripper moves it the full length of the drum. At this stage the cradle of the coil lift tips up and decants the whole coil on the floor. The cylinders which operate the coil lift and stripper carriage are connected to the pressure system through telescopic piping.

The main electrical gear is by Metropolitan-Vickers Electrical Co., Ltd. The mill motors are supplied with power from a motor-generator set comprising a 3,900-h.p., 500-r.p.m., synchronous motor, taking current at 6·6 kV., driving a 2,200-kW., 600-V. main generator and two reel-booster generators of 456 kW. $\pm$ 338 V. The main mill control is on the Ward-Leonard system. Pre-set and accurate control of strip tension is ensured through the use of "Metadyne" control.

The mill is operated from one of two control tables, which are used alternatively, depending on the direction of rolling, and which show all the data necessary for the working of the mill.

The hydraulic power is supplied at 750 lb./sq.in. for the auxiliaries and 3,300 lb./sq. in. for the roll balance from self-contained pumping and accumulator sets housed in " cellars " below the mill (*see* Fig. 4). These are also to the design and manufacture of Davy and United Engineering Company, Limited.

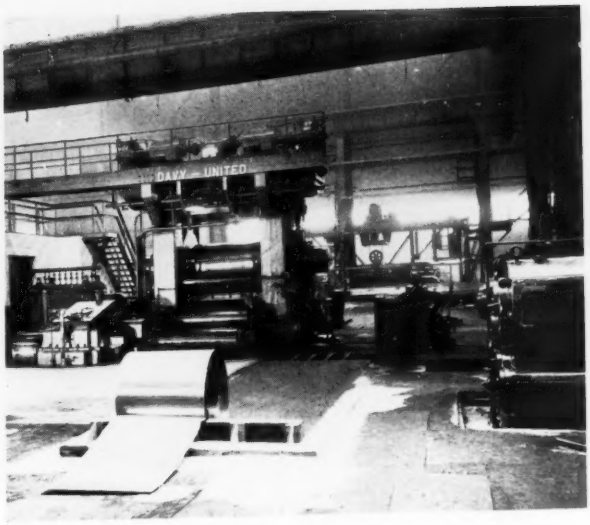


FIG. 1—Delivery side of the mill

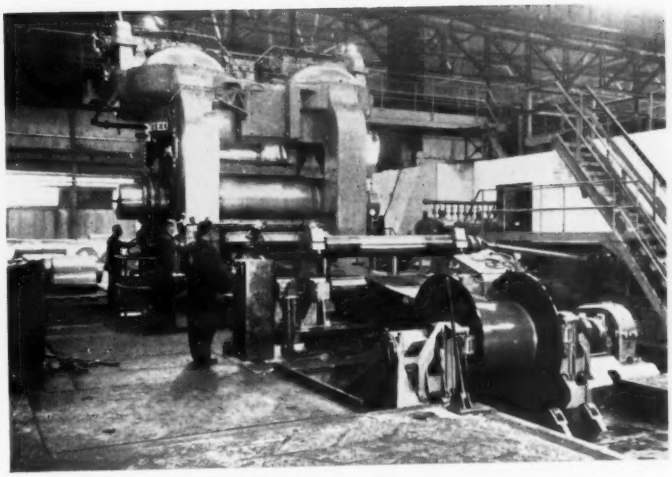


FIG. 2—Entry side of the mill

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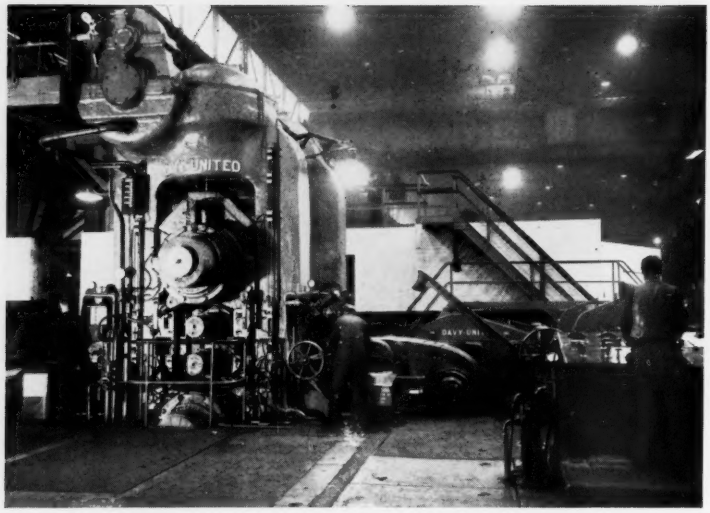

FIG. 3—Roll-change side of the mill, showing control table

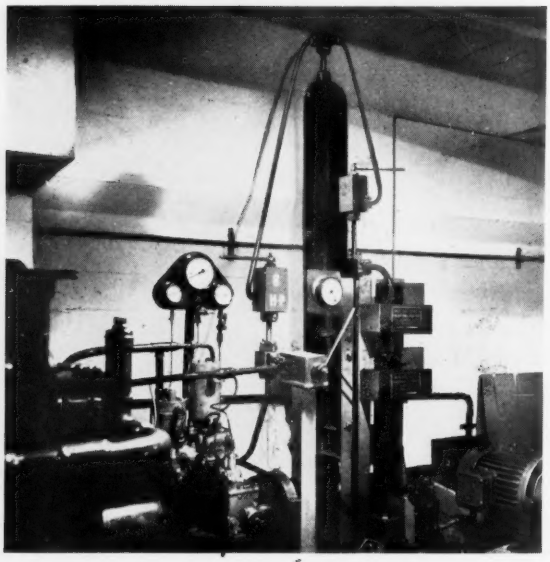

FIG. 4—Air-loaded hydraulic accumulator and pump; 2·2-gal., 3,300-lb./sq. in. working pressure

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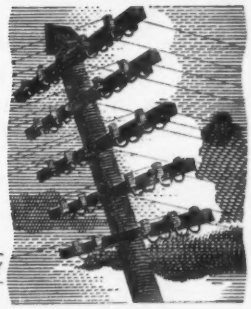
N E W S



ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Secretaryship

In view of the greatly increased activities of the Iron and Steel Institute and of the Institute of Metals, it has been decided by the two Councils that the administration of the work of the two Institutes under a Joint Secretary is no longer practicable and that each Institute should have a separate Secretary. Steps are therefore being taken to implement this decision and, when this process is completed, each Institute will operate with its own Secretary and staff.

The close co-operation between the two Institutes which has been so successful during recent years will not in any way be affected by the change. The two Institutes will continue to share the present premises at 4, Grosvenor Gardens, and the arrangements covering the Joint Library and other matters of common interest will be maintained and strengthened.

Awards

BESSEMER GOLD MEDAL

The Council has decided to award the Bessemer Gold Medal for 1947 to Sir **William J. Larke,** K.B.E., in recognition of his eminent services in the development of research organizations in the iron and steel industry. A brief biographical note and a portrait of Sir William Larke are presented in the frontispiece of this *Journal*.

WILLIAMS PRIZE

The Williams Prize of £100, for 1946, is to be awarded jointly to Dr. **B. Jones**, of Whitehead Iron and Steel Co., Ltd., and Mr. **I. Jenkins**, now of the General Electric Co., Ltd., for their paper on "*New Annealing Plant for Steel Strip in Coils at Whitehead Iron and Steel Co., Ltd., Newport, Mon.*" The paper is published in the *Journal* of the Iron and Steel Institute, 1946, No. II.

ANDREW CARNEGIE SILVER MEDAL

The Andrew Carnegie Silver Medal for 1945 is to be awarded to Dr.-Ing. **M. Balicki,** for his memoir on "*A Study of Work Hardening and Re-annealing of Iron,*" which was published in the *Journal* of the Iron and Steel Institute, 1945, No. I.

An Andrew Carnegie Research Scholarship Grant of £100 was made to Dr. Balicki in 1940, in aid of a research on work-hardening of steel, and it is on the results of this work, carried out at University College, Swansea, that the memoir is based.

ANDREW CARNEGIE GOLD MEDAL

Mr. **H. Morrogh,** of the British Cast Iron Research Association is to receive the 1946 Andrew Carnegie Gold Medal for his memoir on "*Neutralization of Sulphur in Cast Iron by various Alloying Elements,*" which was published in the *Journal* of the Iron and Steel Institute, 1946, No. I.

Mr. Morrogh received a grant of a £100 Carnegie Scholarship in 1940, to aid his research on sulphide inclusions which led to the publication of this memoir.

Symposium on the Hardenability of Steel

The Special Report No. 36, referred to in the News Section of the January issue of the *Journal*, will be published by the Institute in April.

This Report, which will be presented and discussed on 14th May, 1947, in connection with the Annual General Meeting is an account of research on the hardenability of steel, organized by the Ministry of Supply as a result of a visit of British metallurgists to the United States and Canada in 1943.

A full account of investigations into the principles of the end-quench hardenability test is given in the Special Report, of which more complete details of the contents are given on page 452.

Order forms are now being circulated to Members, who are requested to return the completed forms, as soon as possible, if they wish to purchase copies at the reduced Member's rate of 10*s*. each.

Staff Biography, No. 3

Miss Winifred F. Dowd, Chief Accountant, was born in London in 1903. She was trained at Cusack's College, gaining Diplomas in commercial subjects, and later she taught shorthand and bookkeeping at the College for a short period.

Vandyk

On the death of her father, who had been on the staff of the Iron and Steel Institute for nearly twenty-five years, Miss Dowd joined the staff of the Institute, as an assistant in the Accounts Department.

Having gained experience in the business side of the organization, Miss Dowd was then transferred to the Secretarial and Editorial Departments, working under the late Secretary, Mr. George C. Lloyd, and also with Mr. L. P. Sidney and Mr. Alan E. Chattin.

In 1930 Miss Dowd was appointed Accountant and her duties included also the control of the Membership Register and the compilation of Membership Records. She has also been responsible for keeping the Accounts of the Institute of Metals since January 1945, and her services were used in a similar capacity when the Institution of Metallurgists was formed in October, 1945.

During the last war Miss Dowd served as a member of the Women's Voluntary Service and undertook duties as Fire Guard at the Offices of the Institute.

THE IRON AND STEEL ENGINEERS GROUP

The Fourth Meeting of the Iron and Steel Engineers Group will be held at 4, Grosvenor Gardens, London, S.W.1, on Wednesday, 11th June, 1947.

There will be a morning session from 10.30 A.M. to 12.45 P.M. devoted to the presentation and discussion of two papers on "*Steam Generation and Utilization in Iron and Steel Works.*"

In the afternoon session from 2.15 P.M. to 4.30 P.M. a paper on "*The Application of Gas Turbines to Iron and Steel Works Practice*" will be presented and discussed.

A Buffet Luncheon will be held in the Library of the Institute in connection with the Meeting. Further particulars will be announced later. A circular notice about the Meeting will be sent to all Members of the Group in due course.

THE BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Staff Appointments

Mr. W. C. FAHIE, M.Sc., has resigned from his position of Head of the Instrument Section of the Physics Department of the British Iron and Steel Research Association to take up an appointment in the Ministry of Defence.

To succeed Mr. Fahie, Mr. S. S. CARLISLE, M.SC., M.A., a Graduate of Queens University, Belfast (1940), has been appointed Head of the Instrument Section. Mr. Carlisle joined the Association in October, 1946, following four years of service in the Experimental Department, H.M.S. Excellent, Portsmouth, on the staff of the Director of Scientific Research, Admiralty. There he was largely responsible for the design and development of certain special recording instruments and electro-mechanical control systems associated with modern gunnery and radar techniques.

NEWS OF MEMBERS

Recent Appointments

▸ Mr. H. W. SECKER has been appointed Chairman and Managing Director of the Birchley Rolling Mills, Ltd., Oldbury, and Mr. A. J. WAINFORD has been appointed a Director of the Company.

▸ Mr. W. E. J. LEWIS has taken up the position of Principal Surveyor for Steel to Lloyds Register of Shipping on the retirement of Mr. LOUIS RIPLEY.

▸ Mr. STANLEY ROBSON is now acting as consultant to the Imperial Smelting Corporation, Ltd., The Zinc Corporation, Ltd., and New Broken Hill Consolidated, Ltd.

▸ Mr. H. A. WAINWRIGHT, M.Met., B.Eng., Assistant General Manager of Birmetals, Ltd., Birmingham, has been elected to the Board of Directors of Birmetals, Ltd., and Birmabright, Ltd.

▸ Mr. J. CLARK has taken up the appointment of Works Engineer at the Redcar Iron and Steel Works of Messrs. Dorman Long and Co., Ltd.

▸ Dr. H. K. HARDY has resigned from the Aluminium Laboratories, Ltd., and has joined the Laboratory staff of the Fulmer Research Institute, Ltd.

▸ Mr. LAWRENCE H. COPE has been appointed a Metallurgist in the Research Department of Messrs. Stewarts and Lloyds, Ltd., Corby.

▸ Mr. I. H. THOMAS has resigned from Messrs. Hall Telephone Accessories (1928), Ltd., to take up an appointment as Metallurgist at the Shell Refining and Marketing Co., Ltd.

▸ Dr. DANIEL S. EPPELSHEIMER has resigned from Metal Hydrides, Ltd., Beverley, Mass., U.S.A., to accept an appointment at the Missouri School of Mines.

▸ Mr. MAURICE J. OLNEY has resigned from the staff of Messrs. Colvilles, Ltd., and has now become Metallurgist at the British Thermostat Co., Ltd.

▸ Mr. P. I. BRITTAIN has been appointed Chemist in charge of the Central Laboratories of the DeHavilland Aircraft Co., Ltd.

▸ Mr. R. C. TUCKER has accepted an appointment as Foundry Manager at the Leeds Piston Ring Co.

▸ Mr. B. E. HOPKINS has resigned from the Birmingham Laboratories of The Mond Nickel Co., Ltd., to take up an appointment in the Metallurgy Division of the National Physical Laboratory.

▸ Mr. HUGH D. WALTON has been appointed to the Lecturing Staff of the Military College of Science, Metallurgy Branch, Shrivenham.

▸ Mr. S. THOMAS has been promoted from General Manager to become a Director of The Lanarkshire Steel Co., Ltd.,

▸ Mr. ANDREW M. BRYAN, formerly Managing Director of Messrs. Shotts Iron Co., Ltd., has now retired.

▸ Mr. C. M. WHITE has been elected President of the Republic Steel Corporation, Cleveland, Ohio, U.S.A.

▸ Dr. F. W. HAYWOOD has been appointed Technical Director of Messrs. Wild-Barfield Electric Furnaces, Ltd.,

▸ Mr. J. S. BOWDEN has resigned from the Ministry of Supply and has taken up an appointment, as Senior Assistant, in the Technical Service and Development Section of the Tin Research Institute.

▸ Mr. R. BENNETT has joined the St. Anne's Board Mill Co., Ltd., Bristol.

▸ Mr. E. H. FORS is now in this country with the Davy and United Engineering Company, Limited, Sheffield, having left the Morgan Construction Company, Worcester, Mass., U.S.A.

▸ Mr. E. S. GRUMELL has retired from the staff of the Imperial Chemical Industries, Ltd., but will continue to act in an advisory capacity to the Company. He will also continue to act as Chairman of the Fuel Efficiency Committee of the Ministry of Fuel and Power.

▸ Dr. J. W. JENKIN has been appointed General Manager of the newly formed Department of Development and Research of Tube Investments, Ltd., Birmingham.

▸ Mr. H. HUMPHRIES has been appointed a Director of Messrs. Hadfields Ltd., East Hecla Works, Sheffield.

▸ Dr. I. G. SLATER, formerly Officer-in-Charge, Admiralty Central Metallurgical Laboratory, has been appointed Scientific Adviser, British Admiralty Delegation, Washington, U.S.A.

Elections

▸ Dr. C. H. DESCH, F.R.S., President of the Iron and Steel Institute, has been elected an Honorary Member of the American Institute of Mining and Metallurgical Engineers.

▸ Mr. T. F. RUSSELL has been elected President, and Mr. P. JUBB Honorary Secretary, of the Sheffield Metallurgical Association.

▸ Mr. V. C. FAULKNER has been nominated as President-Elect of the Comité International des Associations Techniques de Fonderie, of which he was a Founder in 1926, when President of the Institute of British Foundrymen.

▸ Dr. W. F. CHUBB has been elected a Fellow of the Royal Institute of Chemistry.

Honours

The President and Members of the Council offer their congratulations to :

▸ Mr. A. E. LEEK, General Manager and Director of the Distington Engineering Co., Ltd., who was awarded the M.B.E. in the New Year's Honours List.

▸ Mr. J. A. MCWILLIAM of Messrs. Firth-Vickers Stainless Steels, Ltd., who has been awarded the Hatfield Prize for the best paper read to the Sheffield Metallurgical Association in 1946.

MEMOIRS

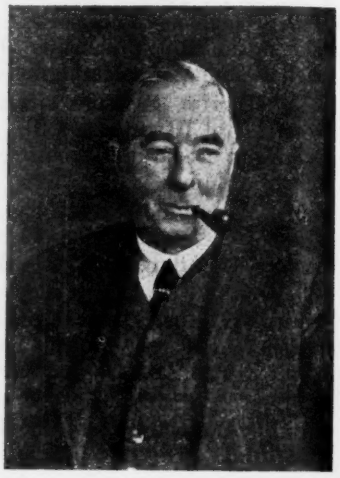
Dr. James William Donaldson, Honorary Member of Council, died on 12th January, 1947, at the age of sixty. He was educated at Falkirk High School and Edinburgh University and, after serving an apprenticeship in the laboratory of the Carron Iron Company, received further training at the Royal Technical College, Glasgow. He served as chief assistant chemist at Messrs. D. Colville and Sons, Ltd., Motherwell, from 1912 to 1914; head of the research laboratory of Messrs. William Beardmore and Co., Ltd., Glasgow, from 1915 to 1918; and from 1918 until his death he was chemist and metallurgist of Scotts' Shipbuilding and Engineering Company, Ltd., Greenock.

In 1914 he was awarded a Carnegie Scholarship by the Iron and Steel Institute, and the results of his research were published in 1916 as a Carnegie Scholarship Memoir, entitled "*Gases Occluded in Alloy Steels.*" (Carnegie Scholarship Memoirs, 1916, vii, p. 41). Later he became a prolific writer of papers for the technical societies and journals on the results of his original research on foundry practice, the properties of cast iron, non-ferrous metallurgy, lubrication, and marine corrosion. His paper on "*The Thermal Conductivity of Wrought Iron, Steel, Malleable Cast Iron, and Cast Iron,*" was read before the Iron and Steel Institute (*see* Journ. I. and S. I., 1933, No. II, p. 255).

Dr. Donaldson was a very active member of the Institute of British Foundrymen. His numerous contributions to its Proceedings were recognized by the award in 1928 of the Oliver Stubbs' Medal, and by his election as a Member of the Council of that Institute from 1929 to 1941. Likewise he read many papers before the West of Scotland Iron and Steel Institute and, in 1945, was elected its President. Another distinction bestowed upon him was his election to be Member of the Council of the Institute of Metals from 1939 to 1943. He was an Associate of the Royal Institute of Chemistry and also Fellow of the Institution of Metallurgists.

He became a Member of the Iron and Steel Institute in 1915.

Mr. James Foster-Smith, Managing Director of The Wellman Smith Owen Engineering Corporation Limited, died at his home in London on 24th January, 1947, at the age of sixty-six. His death, after a comparatively short illness, deprives British industry of a very capable leader who directed the affairs of the Wellman Corporation for more than twenty-five years.



Mr. Foster-Smith was a Member of the Institution of Mechanical Engineers, and his widespread interests included membership of the Council of the British Engineers' Association and other kindred Associations: he was also a Liveryman of the Girdlers' Company and the Founders' Company in the City of London, of which he was a Freeman. He was elected a Member of the Iron and Steel Institute in 1910.

From his early youth his main interest was engineering, and after completing his technical education at Glasgow University he entered the service of the Glasgow Crane and Hoist Company, later joining the staff of Messrs. Appleby and Company, crane builders.

His first association with the Wellman Companies was in 1904 when he was appointed Chief Engineer of Messrs. Wellman Seaver and Head Ltd. and was engaged on the early development of steelworks handling machinery in this country. In 1912 he entered the service of Messrs. Babcock and Wilcox Limited as Engineer Manager of the Crane Department. On resigning this position after a few years, he formed The James Smith Hoisting Machinery Company Limited. In 1919 his Company joined forces with Messrs. Wellman Seaver and Head Ltd., and The Wellman Smith Owen

Engineering Corporation Limited emerged with Mr. Foster-Smith in the chief executive position as Managing Director. From that time his energies were devoted to the development of the Corporation's business and the improvement of British practice in the design and construction of steelworks machinery and metallurgical furnace plant. It was largely through his enterprise and foresight that the design and manufacture of seamless steel tube-making plant was brought to a high degree of perfection in this country. Out of this process new shell-making methods emerged which proved of immense value to Britain in the war.

Dr. A. H. Jay. The death of Dr. Alfred Hartley Jay at the early age of 40, as a result of head injuries received in a car accident on February 22nd, 1947, is a great loss to the world of science.

Dr. Jay possessed a rare combination of gifts. Although a pure scientist of very high standing he could talk to anyone in the steelworks in their own language. Whilst his own peculiar field was X-ray analysis, he made an equal contribution in the field of refractories as well as notable contributions to various aspects of metallurgical work. During the few years prior to his death, he went a long way towards placing the understanding of steelmaking on a simple scientific basis and was obtaining particularly exciting results on the fundamentals of creep strength at the time of the accident.

Dr. Jay won his way to Nelson Secondary School and thence to Manchester University by means of scholarships, graduating in the Honours Physics school in 1928. He was awarded a two-year grant by the Department of Scientific and Industrial Research, and worked with Dr. A. J. Bradley in the Laboratories of Professor W. L. Bragg. He was awarded his M.Sc. for a thesis presented in 1931, and then carried out a further two years work, receiving his degree of Doctor of Philosophy in 1933. During the latter period he held the Priestley Scholarship and the Beyer Fellowship awarded by Manchester University, as well as a research grant from the Metropolitan-Vickers Electrical Co., Ltd., Manchester. The period was a very fruitful one, major contributions being made to the technique of X-ray analysis and standards of accuracy set up which have achieved an international status.

In October 1933, Dr. Jay joined the Central Research Department of the United Steel Companies, Limited, where his work was of outstanding quality. He shed new light on every problem he handled and gave valuable assistance to many workers outside his own immediate field. Not content with the available equipment for the measurement of spacings on films, he designed his own apparatus and followed it with a very simple and practical form of photometer. He was continually improving his own technique, and his masterly ability in sorting out the constituents present in the various complex materials he received for examination was a source of admiration to many experts in his field.

His superb technique was well illustrated by his work on the stabilization of dolomite bricks and the constitution of sea-water magnesia. It was his X-ray analysis of a number of "perished" dolomite bricks that gave the clue as to the real cause of failure and pointed the way to the stabilized dolomite brick as we know it today. His initial X-ray survey of the field of refractories has long since been accepted as a standard. His achievement in the field of fireclay refractories is still scarcely appreciated. Again and again he was able to give information regarding the firing treatment of fireclay bricks that was unobtainable by any other method.

During the last few years he carried out exceptionally fine work on the constitution of steelmaking slags, though much of this work was not published in detail. In all this work he continually strove to relate such practical questions as the difference between "bad" and "good" slags to fundamental mineralogical constitution and rarely failed to simplify the problem, even in the eyes of the practical man. His discovery that so-called liquid basic slag contained a considerable amount of solid material is just one example of the light which he threw on steelmaking processes.

Many technical papers were written by Dr. Jay, both during his term at the University and during his service with the United Steel Companies, Limited. In addition he gave numerous lectures to scientific and technical societies.

Dr. Jay leaves a widow and a young daughter.

J. H. C.

CONTRIBUTORS TO THE JOURNAL

H. Morrogh—Research Manager at the British Cast Iron Research Association has, at the age of 29, spent almost half his life in the service of the Association. He joined the Metallography Department in 1934 and two years later was appointed Head of the Department. His early work, in this capacity, was devoted to improvements in the technique of the preparation of microspecimens, with such evident success that microspecimens prepared in the Laboratories of the Association set a recognized high standard throughout the country.

In 1942, Mr. Morrogh was promoted to Senior Research Officer and was responsible for the planning and supervision of the whole of the metallurgical programme of the Association. He was appointed to his present position in 1946 and his work now covers the whole of the research programme.

For ten years Mr. Morrogh's work has been related to the study of cast irons and he has written many papers giving the results of his investigations.

Mr. Morrogh received an Andrew Carnegie Scholarship in 1940 and, as announced above, he is now to receive the 1946 Andrew Carnegie Gold Medal.

A. H. Leckie, Ph.D.—Deputy Head of the Steelmaking Division of the British Iron and Steel Research Association. Dr. Leckie studied chemistry under Professors Donnan and Ingold at University College, London, spending some years on research on the Raman effect in molecular spectroscopy. In 1937 he joined the Technical Department of the British Iron and Steel Federation (Iron and Steel Industrial Research Council), engaged on research into iron-and-steel plant design with special reference to the study of thermal and aerodynamic conditions on open-hearth furnaces. When the British Iron and Steel Research Association took over the work of the Iron and Steel Industrial Research Council, Dr. Leckie was appointed to the staff of the Steelmaking Division, became Deputy Head of that Division, and is in charge of researches on furnace design and operation.

William John Williams, B.Sc.—Metallurgist on the Research Staff of the British Cast Iron Research Association, was born at Ystradgynlais in 1921. He was educated at the local Secondary School and, matriculating in 1937, he obtained a Central Welsh Board Higher School Certificate in 1939. Mr. Williams entered Swansea University College in 1939 and received the degree of B.Sc. in metallurgy in 1942. During vacation courses he gained practical experience at local steelworks and rolling mills and also at copper smelting and refining works in Swansea.

In 1942, Mr. Williams joined Messrs. Morris Motors Ltd., Coventry, and was engaged in the Foundry Department on the production and casting of electric-furnace-melted manganese steel, and in research work on core-oil substitutes. He joined the staff of the B.C.I.R.A. as metallurgist in 1943 and has since worked on researches into the formation of graphite during solidification of cast iron as well as on other factors relating to the production and properties of cast iron.

Mr. Williams is an Associate Member of the Institute of British Foundrymen.

C. Cartlidge—Assistant Melting Shop Manager of the Shelton Iron, Steel and Coal Co., Ltd. After a period at a small ironworks in the Stoke-on-Trent area, Mr. Cartlidge joined his present Company in 1920 and was closely associated with the pioneer work of Mr. F. L. Robertson in extending the use of instruments for open-hearth furnace control. In conjunction with the instrument manufacturers, he played a leading part in the development of a reliable method for the continuous recording of roof temperatures.

G. Fenton, B.Sc.—Member of the staff of the Steelmaking Division of the British Iron and Steel Research Association. He obtained a First-Class Honours Degree in Metallurgy at Leeds University and, since joining the Association in 1945, has been engaged in research on open-hearth furnaces, principally at the experimental furnace, Shelton.

J. F. Allen, B.Sc.—Member of the staff of the Steelmaking Division of the British Iron and Steel Research Association. Mr. Allen studied Metallurgy and Fuel at Sheffield University and, after graduating, took up a position with the Fuel Department of Messrs. Steel, Peech and Tozer. He gained experience of open-hearth furnaces in that Company's melting shops, and joined the Association's staff in 1945 to take charge of the work in the experimental open-hearth furnace at the works of the Shelton Iron, Steel and Coal Co., Ltd.

J. R. Hall—Melting Shop Manager of the Shelton Iron, Steel and Coal Co., Ltd. Mr. Hall received his early training at the North Staffordshire Technical College, Stoke-on-Trent, and joined his present Company in 1920 as a chemist in the Coke Ovens Laboratory. He became Chief Chemist of that Laboratory and then transferred to the Steel Works, being appointed to his present position in 1940. Mr. Hall has continued the work of his predecessor, Mr. F. L. Robertson, on the strict control of open-hearth furnaces according to instrument readings, particularly the use of roof-temperature measurement in conjunction with gas-and-air flow records. He is now actively concerned with the installation of similar methods of control in furnaces which are being converted to oil firing.

FOURTH EMPIRE MINING AND METALLURGICAL CONGRESS, LONDON, 1948

Preliminary Announcement

Arrangements are being made to hold the Fourth Empire Mining and Metallurgical Congress in London, from 14th to 18th June, 1948.

An Organizing Committee has been set up by the Empire Council of Mining and Metallurgical Institutions, of which the Institute is a constituent. Sir Henry Tizard, K.C.B., A.F.C., F.R.S., has accepted the office of President of the Congress, and Sir Clive Baillieu, K.B.E., C.M.G. (Chairman of the Empire Council) that of Treasurer. Colonel G. M. Carrie, O.B.E., R.C.A., M.C.I.M., is Chairman of the Organizing Committee, and the Secretaries of the Institution of Mining Engineers and the Institution of Mining and Metallurgy are Joint Secretaries of the Congress.

The Congress will consist of seven half-day sessions for the discussion of papers on Economic Geology, Geophysical Prospecting, Mining (coal, metalliferous, and non-metallic), Iron and Steel, Non-Ferrous Metallurgy, and possibly Petroleum. During the week a Banquet and other social functions will take place, and a number of visits to places of interest in and near London will be arranged for ladies accompanying Congress members. For the week following the Congress, visits to mines, works, and places of interest in England, Scotland, and Wales are being arranged.

Members of the Institute who wish to submit papers for discussion at the Congress are requested to communicate with the Secretary of the Institute, and all those who wish to receive further details when available, are asked to notify the General Secretary, Empire Mining and Metallurgical Congress, Salisbury House, Finsbury Circus, London, E.C.2.

X-RAY CRYSTALLOGRAPHY

Design and Production of X-ray Tubes for Crystallographic Purposes

The X-ray Analysis Group of the Institute of Physics has recently established an X-ray Tube Panel, under the chairmanship of Dr. A. Taylor, F.Inst.P., F.I.M. The Panel will review existing and projected X-ray tubes and X-ray tube production in this country, and where supply or quality is not satisfactory, specifications will be prepared for the type of tube concerned and production or improvement will be encouraged by appropriate means.

To enable the review to be as complete as possible a questionnaire is being sent to all members of the X-ray Analysis Group of the Institute of Physics and to certain other persons known to be interested in crystallographic X-ray tubes, but there may be workers in this field who are not included in the lists available.

Anyone interested in X-ray tubes for crystallographic purposes who is not a member of the X-ray Analysis Group, and who wishes to ensure that his views and requirements are considered, is invited to communicate with the Honorary Secretary of the Panel, Mr. J. N. Kellar, A.Inst.P., Crystallographic Laboratory, Cavendish Laboratory, Cambridge.

OBITUARY

The Council regret to record the death of :

Mr. W. GOLDSBROUGH, Technical Engineer of the Lancashire Steel Corporation, Ltd.

Dr. A. H. JAY, of the Central Research Department of the United Steel Companies, Ltd.

An appreciation of Dr. Jay, by his colleague Dr. J. H. Chesters, is given on page 449.

TRANSLATION SERVICE

Since the announcement made in the February issue of the *Journal* (*see* page 296), further translations have been put in hand and the following are now available or in course of preparation :

TRANSLATIONS AVAILABLE

No. 301 (Swedish). R. TOSTERUD and E. STENFORS : "Investigation of the Wear of Rolls during Hot-Rolling of Wire and Strip." (*Jernkontorets Annaler*, 1946, vol. 130, No. 6, pp. 213–238).

No. 302 (German). W. LUEG and A. POMP : "Tests to Determine and Predict the Drawing Force in the Bright-Drawing of Steel." (*Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, 1944, vol. 27, No. 4, pp. 43–52).

No. 303 (German). K. P. HARTEN and H. H. VOSGERAU : "The Present Position and Direction of the German Iron and Steel Industry." (*Stahl und Eisen*, 1947, vol. 66–67, Jan. 2, pp. 2–10).

No. 304 (Swedish). G. PHRAGMÉN and Bo. KALLING : "The Effect of Temperature on the Equilibrium between Carbon and Oxygen in Molten Steel and in a Gas Phase." (*Jernkontorets Annaler*, 1939, vol. 123, No. 5, pp. 199–218).

No. 305 (German). H. UNCKEL : "Phenomena in the Pressing of Metal Powders." *Archiv für das Eisenhüttenwesen*, 1945, vol. 18, Jan.–Feb., pp. 161–167).

TRANSLATIONS IN COURSE OF PREPARATION

(Russian). V. A. MOJAROV : "The Use of Oxygen for Steelmaking in Reverberatory Furnaces." (*Kislorod*, 1946, No. 1, pp. 1–14).

(Swedish). S. MÖRTSELL : "Rationalization in Swedish Iron-Ore Dressing." (*Jernkontorets Annaler*, 1946, vol. 146, No. 9, pp. 369–460).

CHARGES FOR COPIES OF TRANSLATIONS : For the above translations a charge will be made of 10*s*. for the first copy and 5*s*. for each additional copy of the same translation. Requests for translations should be accompanied by a remittance. The above translations are not available on loan from the Joint Library. Members can, however, borrow a translation of the following article from the Joint Library :

(German). H. SCHMITZ : "The Applicability of Aluminium–Magnesium–Silicon Alloys for Killing Steel" (*Stahl und Eisen*, 1947, vol. 66–67, Jan. 2, p. 26.)

TRANSLATIONS PREPARED AT MEMBERS' REQUEST :—Members requiring translations of foreign technical papers are invited to communicate with the Secretary, who will ascertain whether the translations can be prepared for inclusion in the series.

SYMPOSIUM ON THE HARDENABILITY OF STEEL

Special Report No. 36

This Symposium, which was organized by the Technical Advisory Committee (Ministry of Supply) and the Iron and Steel Institute, is in ten Sections and comprises 430 pages. It deals with the following aspects of hardenability and the Jominy end-quench hardenability test :

Effect of changing the test conditions

Reproducibility of hardenability curves

Standard test procedure

Theory of the test

Changes in the surface temperature of the specimen

Correlation of Jominy test data with data from other tests

Variation in the hardenability of heat-treated En steels

Effect of variations in composition

Variation in hardenability between casts of the same steel

Hardenability distribution in billets

Jominy hardenability and mechanical properties of En steels

Effect of alloying elements and boron on hardenability

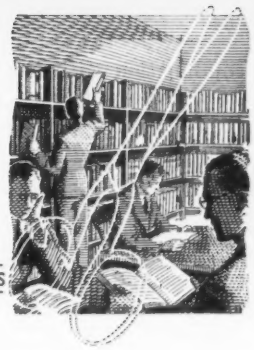
Hardenability of a chromium-molybdenum steel

Information obtainable from hardenability curves

Suggestions for future research

Abtracts of the papers in this Report will be found on pages 472 - 475 of this *Journal*

ABSTRACTS OF CURRENT LITERATURE *and* BOOK REVIEWS



IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

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ORES—MINING AND TREATMENT

Mining and Concentrating in the U.S.A. Part I Mining. S. Dalhammar. (Teknisk Tidskrift, 1947, vol. 77, Jan. 11, pp. 33–38). (In Swedish). The author reports on a visit to the United States in 1946 to study American mining equipment and methods. Data on power consumption, output and costs are given and comparisons are made with Swedish practice.—R. A. R.

Mining and Concentrating in the U.S.A. Part II Concentrating. F. Mogensen. (Teknisk Tidskrift, 1947, vol. 77, Jan. 11, pp. 38–42). (In Swedish). American methods of concentrating ores are reviewed. The great value of pilot-plant investigations is stressed.—R. A. R.

Mining and Concentrating in the U.S.A. Part III The Sink and Float Process in the U.S.A. L. E. Berglund. (Teknisk Tidskrift, 1947, vol. 77, Jan. 11, pp. 42–44). The principles of heavy media separation and some of the equipment used for it in the United States are discussed. The author believes it could be applied with advantage to the concentration of several Swedish ores.—R. A. R.

The Principles Underlying the Sintering of Iron Ores. R. Hay and J. M. McLeod. (Journal of the West of Scotland Iron and Steel Institute, 1944–45, Part VI). The functions and characteristics of iron-ore sinters are discussed. Neustaetter has stated that more of the coke used in the blast-furnace is required for the provision of voids than for thermal or chemical purposes, and the use of sinter should eliminate this waste of coke, but experience has shown that, when the quantity of sinter in the burden exceeds about 35%, the poor reducibility of the sinter begins to outweigh its advantages.

With the object of producing a more easily reducible sinter whilst retaining the strength of the material, a study was made of the mechanism of formation, the micro- and macro-structures, and the composition and reducibility of various classes of sinters. Lime was found to be an important constituent of the mix, but, as the presence of FeO reduced the amount of silica available for

combination with the lime, the optimum lime/silica ratio depended upon the carbon content. In the sintering of ores high in lime the carbon content of the mix must therefore be kept at a minimum. In general, added lime and a low carbon content were found to be beneficial.

The effects of the sintering temperature, ramming of the mix, and particle size of the mix components are also discussed. From the tests on industrial sinters and natural ores it was concluded that great improvement in sinter-plant technique is possible, and practical suggestions for the production of better sinters are given.

The Recovery of Mineral Products by Flotation from the Waste of Iron-Ore Concentration Plants. Part II. G. G. Bring. (Jernkontorets Annaler, 1946, vol. 130, No. 11, pp. 605–647). (In Swedish). Investigations of the factors which control the flotation of hematite, mica, and lime, when oleic acid and soaps are used as collectors, are described. The flotation of apatite was dealt with in Part I (*see* Journ. I. and S. I., 1946, No. II, p. 17A).

Selective flotation of the above-mentioned minerals is possible if the hydrogen-ion concentration is kept under close control and sodium silicate is used as a depressant. Optimum *p*H for oleic acid separation was between 7 and 9 for apatite, between 5 and 6 for hematite, and between 10 and 11 for mica. Apatite concentrates with 30–35% of P_2O_5 have been produced on a large scale from mill tailings during the war and have served as raw material for the manufacture of fertilizers. When flotation is used for the recovery of hematite the difficulty arises that apatite floats together with the iron mineral, even if sodium silicate is added and a concentrate high in phosphorus is produced. This difficulty was overcome by using both sodium silicate and fluosilisic acid as depressants for the apatite. It was thus possible to obtain concentrates with about 65% of iron and less than 0.01% of phosphorus. Tailings containing lime are preferably floated with soaps at a high *p*H-value, whereby sulphides are held back by cyanide and the silica minerals by sodium silicate. Sometimes the lime contains manganese, and the concentrates are then intended to be used for metallurgical purposes. Concentrates from tailings high in silica have been obtained with SiO_2 as low as 2% and CaO up to 49%.

Engineering Problems in the Preparation of Ores for Blast-Furnaces. D. C. Hendry. (Journal of the Iron and Steel Institute, 1947, vol. 155, Jan., pp. 121–135). The paper outlines difficulties encountered in the crushing, screening, and blending of ore for the blast-furnace, with particular reference to Northamptonshire and Lincolnshire ores, and describes the types of equipment in use for these operations, noting the advantages and disadvantages of each type and the precautions necessary to ensure smooth operation.

A critical examination is made of normal sinter plant layout and equipment, and suggestions are given for improvements to raise the quality of sinter produced. Discussion of modern British plants occupies most of the paper, but one of the most modern ore-preparation plants—that of the Hütte Braunschweig, Germany, is also described. One necessity which emerges from a comparative study of different plants is a much closer continuous collaboration between designer and operator, to ensure that improvements are immediately taken up and defects eradicated.

FUEL—PREPARATION, PROPERTIES, AND USES

The Theory of Moving Sources of Heat and Its Application to Metal Treatments. D. Rosenthal. (Transactions of the American Society of Mechanical Engineers, 1946, vol. 68, Nov., pp. 849–865). The fundamental principles of the theory of moving sources of heat are outlined briefly. Point, linear, and plane sources of heat are discussed, and equations derived for linear and two- and three-dimensional flow of heat in solids of infinite size and bounded by planes. These solutions are in good agreement with experimental results when applied to welding problems. A single formula has been derived which enables the fairly accurate prediction of the time and rate of cooling for a wide variety of thicknesses of steel, ranges of temperature, and welding conditions. An attempt has also been made to show how the theory can be applied to the rate of extrusion in continuous casting, and the control of flame-hardening and continuous quenching operations.—C. O.

Industrial Fuel. J. Banks. (Royal Institute of Chemistry: Iron and Steel, 1946, vol. 19. Dec., pp. 775–779). Owing to the current coal shortage in Great Britain, the most suitable types of coal cannot always be obtained for processes in the steel industry. Methods for the most effective use of available types of coal for various steelworks' requirements are outlined and examples given. The chemical control of combustion is also discussed.—J. R.

Research on Coal Utilisation. R. W. Foot. (British Coal Utilization Research Association: Engineering, 1947, vol. 163, Jan. 3, p. 7). The

scope of the work carried out and in progress at the research station of the British Coal Utilization Research Association is reviewed.—R. A. R.

Combustion Safety Devices. W. G. Scharing. (Industrial Heating, 1946, vol. 13, Nov., pp. 1770–1778, 1895–1897). The development of safety devices for use with gas flames is briefly reviewed. The most modern methods for cutting off the gas supply in the event of the extinction of the flame involve the electronic amplification of a small electric current which is passed through the ionized envelope of the flame. Interruption of the current causes a relay to shut off the gas supply. Examples are shown of the use of such a safety device.—C. O.

The Influence of the Moisture Content of Coals on Coke-Oven Operation. (British Coking Industry Association, Fuel Efficiency Bulletin No. 2 : Ministry of Fuel and Power, Fuel Efficiency Committee Conference on "Fuel and the Future", London, Oct., 1946). Recent British, American, and German work on the effect of the moisture content of coals on the operation of coke-ovens is summarized. Among the conclusions drawn from the available data are : (1) Up to 9% of moisture does not cause any great increase in fuel-gas requirements ; (2) low moisture content, 4% or less, frequently lead to charging difficulties.—C. O.

Krupp-Lurgi Low-Temperature Carbonization Plant of Fried. Krupp A.G., at Wanne-Eickel near Bochum. (Combined Intelligence Objectives Sub-Committee, 1946, File No. XXXI–30 : H.M. Stationery Office). The construction and operation of a Krupp-Lurgi low-temperature carbonization plant is described.—R. A. R.

Krupp-Lurgi Low-Temperature Carbonization Plant at Krupp Treibstoffwerke, Wanne-Eickel. (British Intelligence Objectives Sub-Committee, 1946, Final Report No. 817 : H.M. Stationery Office). Corrections and some additions are made to the previous report on this plant (*see* preceding abstract).—R. A. R.

Dry-Hot Cleaning of Blast Furnace Gas—A Revolution in Gas Cleaning. F. E. Kling. (Blast Furnace and Steel Plant, 1946, vol. 34, Oct., pp. 1257–1264). The development of methods of cleaning blast-furnace gas is reviewed. A process has been evolved in which the hot gas is filtered through cloths of glass fibre and asbestos. During tests carried out for one year at a modern blast-furnace, the average dust content of gas cleaned by this method was found to be 0·001 grains/cu. ft., which compares very favourably with the results of electrostatic cleaning.—C. O.

The Gas Balance in Iron and Steel Works. K. Rummel. (Stahl und Eisen, 1947, vol. 67, Jan. 2, pp. 19–23). The balancing of blast-furnace gas supply and demand in integrated iron and steel works is discussed. The limits within which gas production can be adjusted to the demand by changes in the blowing are pointed out and the economic aspects of increasing the normal quantity of coke in the burden by 10% so as to produce more gas are considered.—R. A. R.

Town's Gas for Drying by Radiant Heat. R. F. Hayman. (Ministry of Fuel and Power, Fuel Efficiency Committee Conference on "Fuel and the Future", London, Oct., 1946). The general applications of radiant heat to the finishing of metallic and other surfaces are described, with particular reference to town's gas as the original source of heat.—C. O.

Improvements to Gas Producers. G. Neumann and O. Weber. (Stahl und Eisen, 1947, vol. 67, Jan. 2, pp. 23–26). Several mechanical improvements to gas-producer equipment are described.—R. A. R.

Progress in Hydrogenation of Coal and Tar. K. Gordon. (Institute of Fuel : Chemical Age, 1946, vol. 55, Dec. 21, pp. 761-770 ; Dec. 28, pp. 795-804). A detailed account is given of technical and chemical developments in the hydrogenation of coal and tar at the Billingham plant of Imperial Chemical Industries since 1935, and comparisons are made with the progress made in Germany as revealed in recent investigation reports.—C. O.

The Case for Oil Firing. F. J. Erroll. (British Steelmaker, 1946, vol. 12, Dec., pp. 598–600). The economic advantages of oil-firing in iron and steel works are outlined, and its availability, costs, and application are discussed—J. R.

Liquid Fuel for High Temperature Processes. T. C. Bailey, F. J. Battershill, and R. J. Bressey. (Ministry of Fuel and Power, Fuel Efficiency Committee Conference on "Fuel and the Future," London, Oct., 1946). An outline is given of the physical and thermal properties of fuel oils, and the application of such oils to the production of high temperatures in the metallurgical and ceramic industries.—C. O.

TEMPERATURE MEASUREMENT AND CONTROL

Automatic Control of Heat-Treatment Furnaces. A. Muir. (Iron and Coal Trades Review, 1946, vol. 153, Nov. 29, pp. 973–981). The advantages of controlling the conditions, other than temperature, in metallurgical furnaces are discussed.

Although the initial cost of installing automatic control is considerable, it may be economical in view of the increase effected in the working life of a furnace. Descriptions are given of the use of automatic controls for temperature, furnace pressure, air/gas ratio, combustion, and pressure of combustion air and gas.—C. O.

Boiler Instrumentation. T. H. Carr. (Engineering, 1946, vol. 61, Dec., pp. 201–204, 206). Descriptions are given of the types of instrument required, and their positions in the water and gas circuits, when it is desired to indicate and record what is happening in every part of a steam-raising plant.—C. O.

Thermostat Control in Industry. R. Lloyd. (British Engineering Export Journal, 1946, vol. 29, Dec., pp. 722–728). A number of British thermostats for the control of temperature in manufacturing processes are described and illustrated. —R. E.

Methods of Temperature Measurement of Liquid Iron and Steel. H. T. Clark. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 81–88). Methods used in the United States for measuring liquid steel temperatures are reviewed.—R. A. R.

Principles of the "Black Body Optical Pyrometer" and Instructions for its Use. K. L. Clark. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 88–94). A description is given of the black-body optical pyrometer used by the United States Naval Research Laboratory in their steel foundry. Its calibration and the cost of taking readings are discussed.—R. A. R.

Experience with Open-Tube Immersion Pyrometer. R. B. Sosman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 94–96). Experience gained with an open-tube photoelectric-cell pyrometer for taking liquid-steel temperatures is related. Changes in the pressure of the air which kept the tube clear of fumes and slag had no effect on the readings.—R. A. R.

REFRACTORY MATERIALS

Ladles. W. S. Debenham. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 37–40). After a brief discussion of ladle design, with notes on the Kling transfer ladle, the properties of ladle refractories are dealt with. High density is more readily obtainable in bricks of low refractoriness and these generally have a good life because they are seldom subjected to high temperature and slag for more than an hour at a time.—R. A. R.

Ladle Refractories. M. P. Fedock. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 43–45). A study of the nature and extent of the erosion of the lining of a 15-ton ladle was made. Erosion was most severe at the slag line and less severe within 6–12 in. from the bottom. Between these levels it was even and much less marked. The amount of refractory eroded was 5·25 lb./ton of steel poured; of this, 0·65 lb was attributed to fireclay and nozzle erosion. Trials with a mullite brick lining were unsatisfactory. —R. A. R.

Nozzles and Stopper-Head Properties, Pouring Techniques, and Their Effect on Casting-Pit Performance. L. G. Ekholm and L. D. Hower, jun. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 49-75). A comprehensive account of the design of laddle nozzles and stopper-heads is given with details of the influence of the properties of the refractory materials in question and the manipulating technique on their efficiency and life.—R. A. R.

Sillimanite, Magnesite, and Other Clay-Type Sleeves and Nozzles. M. Corbman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 76-77). The characteristics of sillimanite and magnesite nozzles and sleeve refractories are discussed. Good sleeves were made from a mixture of carbon and clay burned in a reducing atmosphere.—R. A. R.

Brick Linings for Acid Electric Furnaces. C. E. Grigsby. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 158–174). The properties of silica bricks and the design of roofs for acid-lined arc furnaces are discussed in detail. The practice recommended is to heat a new lining to red heat, taking at least 3 hr., and to keep it at above 550°F. before charging. —R. A. R.

Acid Electric Furnace Refractories. C. U. Sellers. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 174–177). Factors affecting the life of the silica brick lining of a 1½-ton Electromelt furnace are discussed.—R. A. R.

Rammed Ganister Linings. G. R. McDaniels. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference,

1945, vol. 3, pp. 179–180). The procedure for making a rammed ganister lining for an Electromelt furnace is described in detail.—R. A. R.

Monolithic Side Walls for Acid Electric Furnaces. C. E. Bales. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 181–184). Experience with a rammed siliceous lining for small Electromelt furnaces is related with details of the method of preparing the material used.—R. A. R.

The Manufacture of Refractories and Information Concerning Their Use in the Iron and Steel Industry of Western Germany. (British Intelligence Objectives Sub-Committee, 1947, F.I.A.T. Final Report No. 432: H.M. Stationery Office). This report covers visits to fourteen German and Austrian plants manufacturing refractory bricks and to two German steelworks. In general, the silica and alumino-silicate refractories have been made by conventional methods in plants which are old and not well laid out. A considerable measure of standardization of the size and shape of many refractory products was achieved and was found worth while. German brickmakers were convinced that Turkish chrome ore produces the best chrome-magnesite brick. The raw magnesite deposits in Southern Austria will soon be worked out if the all-basic open-hearth furnace is generally adopted. The all-basic furnace has met with considerable success in Germany. Experience with carbon-block hearths in blast-furnaces is noted.—R. A. R.

Advances in Refractories Increase Electric Furnace Capacity. F. P. Shonkwiler. (Industrial Heating, 1946, vol. 13, Nov., pp. 1888–1890). A short account is given of increases in the life of electric furnace linings resulting from the use of metal-encased magnesite brick and rammed magnesia.—C. O.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

The Blast-Furnace of Today. Part I—A Review of Current Furnace Engineering. W. R. Brown. (Journal of the Iron and Steel Institute, 1947, vol. 155, Jan., pp. 107–115). The design and layout of blast-furnace plant, equipment and control instruments in the 1946 reconstruction period are discussed with many diagrams and illustrations—R. A. R.

The Blast-Furnace of Today. Part II—A Commentary on Current Furnace Engineering. I. S. Scott-Maxwell. (Journal of the Iron and Steel Institute, 1947, vol. 155, Jan., pp. 115–120). The design of modern blast-furnace plant is critically reviewed. The requirements for economic operation are noted and suggestions for improvement are made.—R. A. R.

Some Notes on Recent American Blast-Furnaces. T. H. Stayman. (Journal of the Iron and Steel Institute, 1947, vol. 155, Jan., pp. 136–140). Constructional details and the operation of four modern American blast-furnaces with hearths from 25 ft. to 28ft. in. dia., producing up to 1500 tons/day are reviewed. The respective advantages of electrical and pneumatic operation for the bells, of mechanically operated and hand-operated bin gates in the stockhouse, and of disintegrators and precipitators for the fine-cleaning of gas are considered.—R. A. R.

PRODUCTION OF STEEL

An Experimental Furnace for the Investigation of Open-Hearth-Furnace Combustion Problems. Part 1. Description of Plant. A. H. Leckie, J. R. Hall, and C. Cartlidge. (Journal of the Iron and Steel Institute, 1947, vol. 155, Mar., pp. 392–405). An experimental furnace specially designed for the investigation of some of the variables affecting open-hearth-furnace fuel efficiency and rate of melting is described. The serious experimental difficulties associated with operation at melting temperatures have been avoided by conducting the experiments at a lower temperature, the heat transferred to the hearth being measured by means of calorimeters. This has permitted the attainment of both a higher degree of accuracy in measurement and greater freedom of experiment than would have been possible if the small-scale melting of steel had been undertaken. The experiments have been designed and regard paid to appropriate similarity criteria so that the results obtained at lower temperatures can be applied to melting practice.

An Experimental Furnace for the Investigation of Open-Hearth-Furnace Combustion Problems. Part II. The Effects of Port Size, Air/Gas Ratio, Furnace Pressure, and Gas Calorific Value. A. H. Leckie, J. F. Allen, and G. Fenton. (Journal of the Iron and Steel Institute, 1947, vol. 155, Mar., pp. 405–422). The effect of air/gas ratio, furnace pressure, calorific value of the gas, and port area in the experimental furnace referred to in the preceding abstract has been studied. In general, the best results are obtained with the theoretical amount of air, high furnace pressures, and high calorific value, provided that the port design is satisfactory. Some of the factors governing correct port design, in particular the relative areas of air and gas ports, are investigated and discussed.

Liquid Fuel for Open Hearth Furnaces. G. R. Bashforth. (British Steelmaker, 1946, vol. 12, Dec., pp. 602–614). As an introduction to a study of the use of liquid fuel in steelworks, the fundamentals of heat transfer in the open-hearth furnace are outlined. The advantages and disadvantages of fuel oil and its characteristics, properties, supply, and storage, are discussed, together with the questions of burner and furnace design and operation. It is concluded that the future of fuel oil is largely a question of economics and national policy.—J. R.

Oxygen in Steelmaking. E. S. Kopecki. (Iron Age, 1946, vol. 158, Nov. 28, pp. 47–50). The literature on the production of cheap oxygen and its use in open-hearth steel furnaces is reviewed.—C. O.

Use of Oxygen in the Open Hearth. G. V. Slottman and F. G. Kerry. (Steel, 1946, vol. 119, Dec. 2, pp. 106–108, 149–152). The possibility of reducing the melting-down period in open-hearth steel furnaces by the use of oxygen-enriched fuel is discussed. The heat input and heat distribution in a furnace can be controlled by using oxygen to determine the character of the flame developed in combustion. Investigations on the effect of oxygen on fuel consumption, waste gases, and refractory life are mentioned briefly.—C. O.

Electric Steel Furnace Transformers. A. G. Arend. (British Steelmaker, 1946, Nov., pp. 553–556). Problems connected with current surges during steelmaking by the electric-arc process are discussed. A compromise is necessary between low current surges and high power factor, and it is recommended that the latter should be between 0·8 and 0·9. Recent trends in the cooling and current regulation of transformers are indicated.—C. O.

Proceedings of the Third Electric Furnace Steel Conference. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, Dec. 4–5, 1945, vol. 3). This Conference, sponsored by the Electric Furnace Steel Committee of the Iron and Steel Division of the American Institute of Mining and Metallurgical Engineers, was held at Cleveland, Ohio. The proceedings were divided into sessions on the following subjects : (1) Ideal conditions for fast melting ; (2) ladles and pouring practice for basic steel ; (3) new methods of temperature control ; (4) economical production of alloy steels ; (5) stainless steel ; (6) furnace and ladle linings for acid steel ; and (7) acid steelmaking operations. In a discussion on melting down turnings containing 1% of carbon and 3·50% of tungsten which were covered with oil, it was stated that the heap was set on fire ; the residue was remelted in a 50-ton basic arc furnace with calcium tungstate and 48% ferrosilicon ; the metallic recovery was 90·8% and the recovery of tungsten 99·50%. Fast melting of 100% scrap charges has been obtained in Germany by oscillating the furnace through 30° on the vertical axis, but this is not necessary in American practice because of the higher power input.

Ladle design is discussed and details of an oval ladle 9 ft. 5 in. × 7 ft. 9 in. at the top and 8 ft. 8 in. × 6 ft. 11 in. at the bottom, with a reduced ferrostatic head, are given.

Stopper rods and sleeves for 8-ton ladles pouring tungsten high-speed steel are described. The cleanest steel was obtained by tapping at 2960–2990° F. after holding for 10 min. in the ladle. The best stopper assembly was a graphite stopper with two graphite and four clay sleeves ; after pouring, these refractories were uncracked, showed little wear, and had fused together into one long stopper. The nozzle was also of graphite.

Discussions on pyrometry brought out the fact that there is greater non-uniformity of temperature in the bath of an electric furnace than in an open-hearth furnace. It was stated with regard to induction-furnace melting that to make alloys and steel high in nickel, magnesia pots were found very unsatisfactory. A rammed lining lasting for 400 heats was made by wetting asbestos cloth with waterglass and fitting it tightly inside the coils ; a 4–5-in. lining of Lannagal was then rammed in the bottom ; an asbestos form was placed in the centre and Lannagal was rammed round it.

Abstracts of the contributions to the Conference follow.—R. A. R.

Preparation of Scrap for Fast Melting. G. A. Barker, jun. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 4–6). The classification of scrap according to analysis and the preparation and charging of scrap for electric furnaces are described.—R. A. R.

Method of Charging and Recharging Small Acid Electric Furnaces, R. H. Frank. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 12–13). Hand-, chute-, and top-charging of electric furnaces of up to 6 tons are described.—R. A. R.

Top *vs.* Machine Charging Basic Electric Furnaces. H. C. Bigge. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 14–16). The top-charging and the machine-charging of large electric furnaces are described. With the former, time is saved by keeping a supply of filled buckets in the scrap yard. More time is required with the latter, but heavier scrap can be charged with less wear on the furnace lining.—R. A. R.

Operations of the Regulex Arc-Furnace Control. C. Wyman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 21–23). A description of the Regulex control and its advantages is given ; it consists of one driving induction motor (220 V., 60 cycles), three generators, and three exciters. It gives high amplification and instantaneous response.—R. A. R.

Power Control on Large Furnaces. H. M. McQuaid. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 24–26). A method by which the operator can easily determine when the furnace is operating at its peak electrical efficiency for the maximum power input to the arc is described.—R. A. R.

Method of Placing Oxidizers, Carburizers and Fluxes in Charge to Effect Fast Melting. J. E. Harrod. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 30–32). The charging practice for a 70-ton Héroult electric furnace is described.—R. A. R.

Method of Placing Oxidizers, Carburizers and Fluxes in Charge. R. J. McCurdy. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 32–34). It is recommended that two or three loads of heavy scrap be placed in the arc furnace before adding ore. With common scrap 50 lb. of ore/ton is sufficient.—R. A. R.

Acid Practice : Oxidizers, Carburizers and Fluxes in Charge. J. Juppenlatz. (American Institute of Mining and Metallurgical Engineers' Electric Furnace Steel Conference, 1945, vol. 3, pp. 34–36). The charging of an acid electric furnace is described with special reference to working a heat down from 1% to 0·25% of carbon.

Ladles. W. S. Debenham. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 37–40). *See* p. 456.

Ladle Refractories. M. P. Fedock. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 43–45). *See* p. 456.

Nozzles and Stopper-Head Properties, Pouring Techniques, and Their Effect on Casting-Pit Performances. L. G. Ekholm and L. D. Hower, jun. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 49–75). *See* p. 456.

Sillimanite, Magnesite, and Other Clay-Type Sleeves and Nozzles. M. Corbman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 76–77). *See* p. 456.

Methods of Temperature Measurement of Liquid Iron and Steel. H. T. Clark. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 81–88). *See* p. 456.

Principles of the "Black Body Optical Pyrometer" and Instructions for Its Use. K. L. Clark. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 88–94). *See* p. 456.

Experience with Open-Tube Immersion Pyrometer. R. B. Sosman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 94–96). *See* p. 456.

Factors in Economic Production of Alloy Steels. L. H. Nelson. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 100–104). The economic aspects of the production of steel in the electric furnace are considered. Special reference is made to the relative merits of providing the required alloying elements in the initial charge, or by later additions.—R. A. R.

Slag Practice and Alloy Additions. R. J. McCurdy. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 105–106). Slag practice in the production of alloy steel is discussed with special reference to the factors affecting the times for flushing-off the slag and for making alloy additions.—R. A. R.

Practice at Carnegie-Illinois Steel Corporation, Duquesne. E. J. Chelius. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 106–107). Electric-furnace practice with one 35-ton and two 70-ton Héroult furnaces is described. When about half the charge has melted oxygen is blown into the bath in cases where the phosphorous has to be very low. The procedure is described in detail. The ram of the charging car is equipped with a mechanical mixer or plate and, after all the charge had melted, this is used to mix the bath thoroughly and obtain uniformity of analysis—R. A. R.

Factors in the Economical Production of Alloy Steel in the Basic Arc Furnace. H. C. Bigge. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 108–111). A detailed log and description of the working of a 28-ton heat in a basic-lined arc furnace to produce a nickel-chromium-molybdenum steel are given.—R. A. R.

Deoxidation in the Electric Arc Furnace. H. C. Bigge. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 112–113). The methods of forming the first and second slags and their purpose in electric furnace practice are discussed.—R. A. R.

Practice at Rotary Electric Steel Co. H. E. Phelps. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 13, pp. 113–114). Double-slag electric-furnace practice at the works of the Rotary Electric Steel Company is described.—R. A. R.

Practice at Carnegie-Illinois Steel Corporation, South Works. J. H. Eisaman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp.114–115). A very high percentage of alloy steel scrap makes up the charge in the electric furnace practice described.—R. A. R.

Practice at Ford Motor Co. W. E. Hart. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, p. 115). A double-slag practice is employed at this plant. To make the first slag 1200 lb. of limestone are added to a 15-ton charge and the second slag consists primarily of 400 lb. of lime and 75 lb. of fluorspar.—R. A. R.

Desulphurization in the Basic Arc Furnace. H. C. Bigge. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 118–120). Desulphurization in a 7-ton basic-lined arc furnace is discussed with special reference to the relation between the final silicon content and the percentage of sulphur removed.—R. A. R.

Single Slag Heats of Tool Steel. A. J. Scheid. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 122–123). In working a heat of tungsten–chromium–vanadium high-speed steel it is found that, after the whole charge is melted, the percentage WO_3 in the metal decreases very rapidly at first, whilst the Cr_2O_3 and V_2O_5 decrease more slowly.—R. A. R.

Stainless Steel Melting Practice at Duquesne. E. J. Chelius. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 125–128). The practice for making 45-ton heats of stainless steel in a basic-lined Héroult furnace is described.—R. A. R.

Practice at American Rolling Mill Company. C. R. FonDersmith. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 130–131). In this description of the practice for making 18/8 stainless steel in the basic-lined electric furnace the importance of separating chromium–nickel steel scrap from straight chromium steel is emphasized.—R. A. R.

Stainless Practice at Atlas Steels Ltd. A. C. Texter. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 131–133). In melting down high-silicon stainless steel it was necessary to charge sufficient ore to oxidize all the silicon and a gassy bath resulted; the gassy condition was corrected by introducing nitrogen.—R. A. R.

Practice at Rustless Iron and Steel Corporation. A. L. Feild. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, p. 133). A process of making stainless steel from chrome ore and stainless-steel scrap in a Héroult furnace is described.—R. A. R.

Practice at Allegheny Ludlum Steel Corporation. W. M. Patterson. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 135–136). The preparation of stainless-steel turnings for charging into electric furnaces is described.—R. A. R.

Stainless Steel Melting and Alloy Recovery at Republic Steel Corporation, Canton District. W. M. Farnsworth. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 139–140).—R. A. R.

Practice at Carnegie-Illinois Steel Corporation, South Works. J. H. Eisaman. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 140–141). Electric furnace practice for making stainless steel in Héroult furnaces is described.—R. A. R.

Ajax Induction Melting of Stainless, Tool and Speciality Steels. C. C. Wilson and C. B. Post. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 142–144). Melting practices for making stainless, high-speed, and high-alloy steels in the Ajax induction furnace in 1200-lb. heats are described.—R. A. R.

Use of Inert Gas to Flush Dissolved Gas from Molten Steel. P. M. Hulme. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 145–151). Experience gained in blowing nitrogen and argon through molten steel to remove dissolved gas is discussed. About 4–5 cu. ft. of gas per ton of steel are required with a flushing time of 5 mins. or less; the hotter the bath, the shorter is the time required. Comparisons are made between flushing aluminium and flushing steel—R. A. R.

Mold and Nozzle Preparation for Stainless Steel. B. A. Roderick. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 154–156). Ingot moulds for stainless steel are prepared by heating them to 180–235° F. and spraying them with a mixture of aluminium paste and an oil. A clay nozzle with a graphite stopper is used for controlling the pouring.—R. A. R.

Brick Linings for Acid Electric Furnaces. C. E. Grigsby. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 158–174). *See* p. 456.

Acid Electric Furnace Refractories. C. U. Sellers. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 174–177). *See* p. 456.

Rammed Ganister Linings. G. R. McDaniels. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 179–180). *See* p. 456.

Monolithic Side Walls for Acid Electric Furnaces. C. E. Bales. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 181–184). *See* p. 457.

Krupp Rings for Roofs. C. W. Briggs. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 183–184). A detailed description is given of the Krupp cast-steel water-cooled rings which are fitted to surround the electrodes where they pass through the roof of the furnace.—R. A. R.

A Comparison of Acid and Basic Electric Practice. I. W. Sharp. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 187–188). The performances of a 3-ton acid and a 9-ton basic electric furnace are compared. Better control of the manganese was obtained in the former, and of the silicon and carbon in the latter. The power consumptions were 525 kWh./ton of charge for the former and 475 kWh./ton for the latter.—R. A. R.

Basic Brick Linings. K. T. Apgar. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 189–190). The procedure for constructing a basic brick lining for electric furnaces is described.—R. A. R.

Bottom-Pour Ladle Practice. C. B. Jenni. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 194–199). The construction of ladle nozzles and stopper assemblies is described and defects in stoppers revealed by X-rays are discussed.—R. A. R.

Bottom-Pour Ladle Linings. T. H. Burke. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 199–201). The lining and drying of ladles and the preparation of stopper assemblies are described.—R. A. R.

Steel Foundry Ladle Linings and Their Repair. J. A. Bowers. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 202–207). The relining of teapot-type ladles is described.—R. A. R.

Acid Ladle Linings. C. H. Welch. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 207–209). The relining of a 6-ton teapot-type ladle, a 1200-lb. ladle, and a small hand ladle, all with acid materials, is described. —R. A. R.

Teapot Ladles and Spouts. E. M. Sarraf. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 209–214). The lining of ladles and ladle spouts with specially shaped bricks and with rammed materials for acid steelmelting practice is described in detail.—R. A. R.

Studies on Susceptibility of Cast Steels to Graphitization. J. J. Kanter. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 218–227). *See* p. 157.—R. A. R.

Graphitization in Cast Steels. A. J. Smith, J. Urban, and J. W. Bolton. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 227–230). *See* Journ. I. and S. I., 1946, No. II, p. 57A).—R. A. R.

Slag Control for the Production of Silicon-Killed Steel in the Acid Electric Furnace. J. B. Caine. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 230–232). The production in an acid electric furnace of coarse-grained silicon-killed steel with a good resistance to graphitization is described.—R. A. R.

Causes and Prevention of Intergranular Fracture in Cast Steel. C. H. Lorig. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 234–240). *See* p. 472.

The Appearance of Rock Candy Fracture. M. F. Hawkes. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 240–244). *See* p. 472.

The Fluidity of Cast Steel—Its Measurement, Its Meaning and Its Value to the Foundryman. H. F. Taylor. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 252–256). *See* p. 462.

Data on Fluidity of Cast Steels. K. L. Clark. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 256–261). *See* p. 462.

Industry Survey of Chromium and Manganese Recoveries. J. F. Robb. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 275–276). Information collected by the Steel Founders' Society of America on the percentage recovery of chromium and manganese from scrap is presented.—R. A. R.

Electrodes—Carbon and Graphite. T. L. Nelson. (American Foundryman, 1946, vol. 10, Oct., pp. 40–48). An account is given of the preparation and physical and mechanical properties of carbon and graphite electrodes. The relative advantages of the two types of electrode and the precautions which must be taken in their use are discussed—C. O.

Arc Furnace Switching. L. W. Long and J. F. Chipman. (Iron and Steel Engineer, 1946, vol. 23, Nov., pp. 53–59). The factors which influence the selection of switch gear for electric arc furnaces are discussed. The majority of the switching devices at present in use are of the oil circuit-breaking type, but in view of the high cost of maintenance with these, the trend is towards the air-blast switches, one of which is described in detail.—C. O.

Electronic Equipment for Induction Melting Alloy Steel. S. R. Durand. (Foundry, 1946, vol. 74, Nov., pp. 118–120). Mercury-arc equipment for changing 3-phase, 60-cycle current into single-phase, 1000-cycle current for an induction furnace with a melting capacity of 650 lb. of alloy steel is briefly described.—C. O.

FOUNDRY PRACTICE

Cupola Melts above Rated Capacity. P. Dwyer. (Foundry, 1946, vol. 74, Dec., pp. 98–101, 233–234). A detailed, illustrated, description is given of the foundry practice of the Forest City Foundries Co., Cleveland, Ohio. Mention is made of a 42-in. dia. cupola, with a normal melting rate of 7 tons/hr., which was operated at the same rate as a 54-in. dia. cupola, *viz.* at a melting rate of 12 tons/hr., by blowing at a pressure of 30 oz. —C. O.

The Hot Blast Cupola. M. Bader. (Foundry, 1946, vol. 74, Dec., pp. 104–107, 252–254). This is an English translation of an article which appeared in Giesserei, 1943, vol. 30, Nov., pp. 241–246 (*see* Journ. I. and S. I., 1946, No. I, p. 19A).

Compressed Air Isn't All Wet. G. B. Comfort. (Foundry, 1946, vol. 74, Dec., pp. 102–103, 201–204). An outline is given of measures which can be taken to reduce the moisture content of compressed air. Suitably arranged mechanical traps are often sufficient to keep the moisture below a desired level; really dry air is obtained by cooling and precipitation of the moisture.—C. O.

Reducing Cast Iron Grain Growth. (Steel, 1946, vol. 119, Nov. 4, pp. 101, 148). A brief account is given of the Metcolizing process for preventing the growth of cast iron caused by oxidation of graphite flakes. The penetration of hot gases into the iron is resisted by a thin coating of aluminium oxide overlying a zone of solid solution of aluminium in iron, produced by heating a metallized aluminium coating at first in the absence and then in the presence of air.—C. O.

Modernizing Malleable Iron Foundries. F. A. Pampel. (Foundry, 1946, vol. 74, Dec., pp. 86–89, 226–232). The various types of equipment which may be used in the mechanization of the

production of malleable iron castings are discussed. Special reference is made to the problems of mould and sand handling.—C. O.

Silicon Pick-Up in Melting Malleable Iron. J. E. Rehder. (American Foundryman, 1946, vol. 10, Oct., pp. 50–54). The physicochemical theory underlying the pick-up of silicon during the melting of iron in the cupola is discussed. The factors which promote an increase of silicon in the iron are: (1) High temperatures (1550° C. or more, for malleable iron); (2) high carbon content, since the reduction of silica is nearly proportional to the square of the carbon content; (3) prolonged heating; and (4) high silica content of refractories or slag.—C. O.

Power Trucks for Foundry Use. F. H. Slade. (Foundry Trade Journal, 1946, vol. 80, Dec. 26, pp. 431–432). The uses and advantages of power trucks for the handling of castings, raw materials, slag, cores, etc., in the foundry are discussed, and a new model is described and illustrated.—J. R.

Mechanized Bathtub Foundry. P. Dwyer. (Foundry, 1946, vol. 74, Nov., pp. 68–73, 156). An illustrated description is given of the casting and enamelling of iron bathtubs in a foundry at Monaca, Pennsylvania. The chief feature of the factory is the straight-line automatic conveyor mechanism.—C. O.

Mechanical Pouring Unit Handles 11 Tons Per Hour. J. R. Shepard and W. J. Olson. (Iron Age, vol. 158, Dec. 5, pp. 77–80). An illustrated description is given of a process developed by the American Brake Shoe Co., New York, which enables 370 brake-shoe moulds to be poured per hour by one operator seated at a push-button control panel.—C. O.

The German Centrifugal Castings Industry with Special Reference to the Production of Cast Iron Pipe, Cylinder Liners and Piston Rings. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 700: H.M. Stationery Office). This report presents a general picture of the German centrifugal castings industry as obtained by visits to 17 foundries. The casting machines were of normal design and known layout. One development was the centrifugal casting in dies of pipe flanges for screwing to double-spigot pipes produced from scrap socket pipes. A long life of the cupola linings was obtained and in some cases the refractory consisted mainly of a mixture of extremely coarse and extremely fine particles.—R. A. R.

The Fluidity of Cast Steel—Its Measurement, Its Meaning and Its Value to the Foundryman. H. F. Taylor. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 252–256). Work on fluidity tests for steel foundries is reviewed, and a spiral fluidity test developed by the United States Naval Research Laboratory is described.—R. A. R.

Data on Fluidity of Cast Steels. K. L. Clark. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 256–261). The effect of manganese, nickel, chromium, and molybdenum on the fluidity of cast steel is discussed.—R. A. R.

Education in the Foundry. J. B. Longmuir. (Institute of British Foundrymen: Foundry Trade Journal, 1947, vol. 81, Jan. 9, pp. 29–35). This article is the conclusion of the description of the training scheme for foundry apprentices, the first part of which appeared in Foundry Trade Journal, 1946, vol. 79, May 16, pp. 55–59 (*see* Journ. I. and S. I., 1946, No. II, p. 22 A).

Cleaning Costs in a Small Gray Iron Foundry. A. E. Grover. (Foundry, 1946, vol. 74, Nov., pp. 83, 188–194). The numerous methods used by the managers of small iron foundries in America to determine the cost of cleaning castings are discussed and compared.—C. O.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Hardness and Hardenability. J. W. Donaldson. (Metal Treatment, 1946, vol. 13, Autumn Issue, pp. 169-181). Recent research work in the United States on hardening and the hardenability of steel is reviewed. The bibliography contains 26 references to the American literature.—C.O.

Sub-Zero Treatment of Steels. G. H. Jackson. (Aircraft Production, 1946, vol. 8, Dec., pp. 588–593). The metallographic considerations involved in the sub-zero-temperature treatment of high-speed and high-alloy steels are discussed. The object of such treatment, an almost completely martensitic structure, can also be attained by a "multiple-tempering" heat-treatment cycle. The author maintains that insufficient practical work has been done to substantiate claims for longer tool life obtained with low-temperature treatment, and that multiple-tempering is still best and safest. The use of low temperatures for alloy steels such as S.82 (0·18% of carbon, 4·0% of nickel, 1·0–1·6% of chromium, plus vanadium, molybdenum, or tungsten) may, however, be advantageous because of the dimensional stability, freedom from stress, and increased hardness of intricate shapes which can be secured.—C.O.

Prepared Atmospheres. C. C. Eeles and M. E. Shriner. (Steel, 1946, vol. 19, Dec. 2, pp. 110–113, 154–156). The controlled furnace atmospheres used in the heat-treatment of iron and steel are classified according to their method of preparation and their use. The action of each of the common industrial gases upon iron and steel is indicated very briefly.—C. O.

Maximum Carbon in Gas Carburizing. G. H. Boss. (Metal Progress, 1946, vol. 50, Oct., pp. 657–658). A brief account is given of experiments on the sintering of iron-powder compacts in natural gas at 2050° and 2200° F. The carbon content of the specimens after treatment at 2050° F. was found to be about 3·3%. It is suggested that chemical affinity rather than a physical force connected with the solution of carbon in austenite is responsible for this very high absorption. The specimens sintered at 2200° F. were hypo-eutectoid in composition, probably because the ingress of carbonaceous gas was prevented by the closing of the voids.—C. O.

Carburizing Treatments. (Steel, 1946, vol. 119, Nov. 18, p. 101). The three carburizing processes in use at the plant of the National Cash Register Co. are briefly described. The carburizing atmosphere consists of a mixture of ammonia, natural gas, and a gas formed by cracking the products of combustion of natural gas.—C. O.

Gas Cyaniding. H. M. Parshall. (Steel, 1946, vol. 119, Dec. 16, pp. 86-88). A brief account is given of the case-hardening of mild steel in gaseous cyanide atmospheres.—C. O.

A New Heat Treating Atmosphere Producer. A. G. Hotchkins and H. M. Webber. (Iron Age, 1946, vol. 158, Nov. 21, pp. 48-55). A new technique for producing an atmosphere such as Neutralene, free from carbon dioxide and moisture, is described, with flow sheets. Considerable reductions in cost and space required have been effected by utilizing heat, which in other producers is wasted, to reactivate the carbon-dioxide-absorbing liquid. Illustrations are given to show the advantages of the Neutralene atmosphere for the heat-treatment of steel.—C. O.

Conveyorized Automatic Heat Treatment. A. F. Holden. (Iron Age, 1946, vol. 158, Nov. 14, pp.70–74). Several types of automatic conveyor mechanisms for use in salt baths are described and illustrated.—C. O.

Automatic Heat Treating Machine. J. R. Morrison. (Metal Progress, 1946, vol. 50, Nov., pp. 952–954). A brief description is given of an automatic pusher-type continuous heat-treating installation. Machined parts are fed into the hardening furnace at one end, and are removed 6½ hr. later at the other end, hardened, quenched, and tempered. There are two furnaces, one for hardening and one for tempering, each with a hearth consisting of a pair of roller rails. The oil-quench tank and washing machine are situated in line between the furnaces.—C. O.

Inspection and Testing for the Heat Treatment Shop. Part 1. Heat Treatment. E. H. Currick. (Australian Institute of Metals: Australasian Engineer, 1946, Oct. 7, pp. 79–82). An account is given of the practice used in carburizing and heat-treating forgings, bars, chains, and axles at the Victorian Railway Workshops, Newport, Australia.—C. O.

Modern Methods for Interrupted Quenching. C. H. Lekberg. (American Gas Association: Industrial Heating, 1946, vol. 13, Oct., pp. 1591–1602). The structural changes which take place during the heat-treatment of a straight-carbon hypo-eutectoid steel are briefly explained as a basis for discussing austempering, isothermal quenching, and martempering.—C. O.

Transporting Quenching Tanks. (Steel, 1946, vol. 119, Nov. 18, pp. 98, 138). A brief description is given of the use of portable quenching tanks in maintaining a continuous flow of parts to and from heat-treatment furnaces. The tanks, mounted on skids, are transported by power-driven platform trucks.—C. O.

FORGING, STAMPING, DRAWING, AND PRESSING

Precision Forging of High Temperature Alloys. C. H. Smith, jun. (Iron Age, 1946, vol. 158, Nov. 28, pp. 42–46). The difficulties experienced in forging heat-resistant alloys are discussed and illustrated by an account of the precision forging of blades for gas turbines.—C. O.

Investigations Concerning Design, Material, Technique, and Production Methods for Valves in Germany. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 910: H.M. Stationery Office). Descriptions are given of the methods of manufacturing valves for automobile engines at three large works in Germany. Special attention is given to the grain flow in the upsetting process.—R. A. R.

Forging Alloy Steel Radial Engine Crankcases. A. J. Pepin and A. L. Rustay. (Metal Progress, 1946, vol. 50, Nov., pp. 1023–1028). The sequence of operations involved in forging radial-engine crankcases from blanks of chromium-nickel-molybdenum steel is described and illustrated.—C. O.

Slide Rule for Calculating Press Size. (Iron Age, 1946, vol. 158, Nov. 28, p. 69). A description is given of a slide rule with which the size of press required for a given drawing or blanking operation may be calculated.—C. O.

Some Drawing Operations on Sheet Metal. H. W. Swift. (Institute of Vitreous Enamellers : Sheet Metal Industries, 1946, vol. 23, Dec., pp. 2365–2376, 2384). Short descriptions are given of some of the pressing operations used in forming articles from steel sheet. Several typical examples are illustrated.—R. A. R.

Drawing Stainless Steel. (Materials and Methods, 1946, vol. 24, Nov., pp. 1168–1169). A short description is given of the " Sol-A-Die " process for drawing stainless steels which was developed between 1939 and 1943 by the Solar Aircraft Company. The dies are so designed, using a flexible laminated beeswax pattern, that the final area of the finished product is reached in the first drawing operation ; the later stages are largely folding rather than drawing operations—C. O.

The Development of Wire-Drawing. C.-O. von Hofsten. (Teknisk Tidskrift, 1946, vol. 76, Aug. 10, pp. 751–758). (In Swedish). A comprehensive review is presented of modern wire-drawing methods with many illustrations of the machinery and equipment used.—R. A. R.

ROLLING-MILL PRACTICE

German Steel Rolling Mill Practice. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 954 : H.M. Stationery Office). This report covers the plant and procedure at four German rolling mills producing : (*a*) High-alloy steel wire and steel for the cutlery trade ; (*b*) drill steel ; (*c*) transformer, dynamo, deep-drawing and tinplate steels ; and (*d*) strip for razor blades. Conventional methods were employed and nothing of particular interest was noted.—R. A. R.

Calculating the Size of Motors for Reversing Mills. A. Geleji. (Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy, 1943, vol. 15, pp. 246–264). Diagrams showing the changes in the revolutions of the rolls per minute in relation to time in reversing-mill operations are presented, and it is shown how to apply these diagrams in designing the mill motors.—R. A. R.

Direct-Current Steel Mill Motor Protection. W. M. Shallcross. (Iron and Steel Engineer, 1946, vol. 23, Nov., pp. 99–101). A detailed description is given of a device for protecting a rolling-mill motor from excessive torque and heating, permitting the maximum utilization without shortening the life of the machine. The control combines a magnetic relay for cutting off the current at a predetermined torque and a thermal overload relay which operates if a moderate overload, insufficient to damage the motor materially, is exceeded.—C. O.

Further Work on the Determination of the Widening in Rolling. E. Cotel. (Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy, 1943, vol. 15, pp. 3–11). The relation between the percentage reduction in thickness and the amount of spread in rolling was dealt with in an earlier paper (*see* p. 308). In this report a new formula is developed for cases of heavy reduction in the 50–80% range.—R. A. R.

Development of Balanced Proportion Roller Bearings for Rolling Mills. P. Haager. (Iron and Steel Engineer, 1946, vol. 23, Nov., pp. 62–72). An account is given of the developments which have been made in the design of bearings for rolling mills as a result of both practical experience and research. A study of the relations between dimensions, roll-neck pressures, and neck rigidity in rolling-mill bearings has led to the design of heavy-duty tapered roller bearings of the " balanced-proportion " type, in which all the bearings in a given line are subjected to equal neck stresses when each is loaded to a given percentage of its rating. The diagrams in the paper show the arrangement of bearings in a 52-in. universal mill, an old type of roller bearing in a 4-high hot-strip mill, the bearings of a 120-in. reversing aluminium mill, applications of "balanced-proportion " bearings, and the design of roll-neck fillets.—C. O.

Metallic Bearings. H. L. Smith. (Iron and Steel Engineer, 1946, vol. 23, Nov., pp. 73–76). The answers submitted by the author to a questionnaire circulated by the Association of Iron and Steel Engineers on the design and operation of metallic bearings are presented.—C. O.

Bearing Seals. L. G. Krug. (Iron and Steel Engineer, 1946, vol. 23, Nov., pp. 77–82). Illustrated descriptions are given of the various type of seals used to prevent the entry of water and scale into the roll-neck bearings of steel mills. The two types of seal in common use are the shaft, in which the sealing element has a sharp edge in constant contact with the shaft so that the surface films of oil, grease, or water are cut, and the end-face or surface type, in which the sealing element is a lapped surface, in tight contact with a lapped surface on the rotating shaft.—C. O.

WELDING AND FLAME-CUTTING

The Flash Welding of Hard-Drawn High Carbon Steel Wire. R. W. Bennett and R. D. Williams. (Welding Journal, 1946, vol. 25, Oct., pp. 658-s–668-s). A process is described for flash welding cold-drawn, high-carbon steel wire with a tensile strength of 220,000 lb./sq. in. so that the resulting joints have 90% of the strength of the parent metal. The cooling of the weld is stopped at 1000° F., the wire being automatically maintained at that temperature in the flash-welding die until the austenite formed in the weld zone has transformed into the relatively soft and ductile pearlite. Joints have been made on a production basis at rates exceeding one per min. —C. O.

Welding of High-Strength Constructional Steels with Ferritic Electrodes. S. A. Herres and P. E. Woodward. (Welding Journal, 1946, vol. 25, Oct., pp. 646-s–658-s). Suitable steels, ferritic electrodes, and welding procedures for the fabrication of structures with minimum yield strengths of 90,000–150,000 lb./sq. in. are discussed and illustrated. A "double **T**" fillet weld specimen is advocated as a representative test for freedom from cracking of parent metal during the welding of restrained joints. Data are presented on the notched-bar impact strength of weld metal and welded joints and on the strength of fillet welds. —C. O.

Welding of Hardenable Steels with High-Alloy (Austenitic) Electrodes. S. A. Herres and A. M. Turkalo. (Welding Journal, 1946, vol. 23, Oct., pp. 669-s–696-s). Data are given on the tensile and impact properties of selected parent metals and austenitic weld deposits, and the mechanical properties of welded joints in various constructional steels are discussed. Welds of good strength and toughness can be obtained without preheating or post-heating in high-strength heat-treated steels by the use of chromium–nickel austenitic electrodes. The most serious defect of such welds is a tendency for failures along the fusion line, which may be reduced by a suitably designed joint. Slow cooling or reheating of austenitic welds above 800° F. causes extreme embrittlement at the fusion zone. The causes and types of welding defects and failures are examined and classified.—C. O.

How to Select Wear-Resisting Alloys for Welding. J. A. Cunningham. (Welding, 1946, vol. 25, Oct., pp. 915–920). The problem of choosing welding electrodes for applying hard surfaces to worn parts is discussed in detail. Some of the factors to be considered are: (1) The severity of abrasion; (2) the type of impact which may be experienced; (3) the possible rewelding of the deposit; (4) the size of the part to be surfaced; (5) the thickness of deposit required; (6) the composition of the base metal; (7) subsequent work-hardening; (8) corrosion; (9) heat-resistance; and (10) subsequent heat-treatment or machining. A table is given showing the compositions, uses, costs, properties, and welding characteristics of a number of hard-facing alloys. —C. O.

Reduction of Residual Welding Stresses by the Use of Austenitic Electrode. F. Jonassen, J. L. Meriam, and E. P. DeGarmo. (Welding Journal, 1946, vol. 25, Sept., pp. 489-s–491-s). Experiments designed to compare the residual welding stresses produced by ferritic and austenitic welding electrodes are reported. The longitudinal residual stresses in welds made by different techniques were always less using austenitic electrodes, but the reduction was not such as to justify the greater expense.—C. O.

Pressure Gas Welding of Alloy Steel Tubing. J. L. Zambrow and R. D. Williams. (Welding Journal, 1946, vol. 25, Oct., pp. 585-s–596-s). A description is given of an investigation into the factors which control the quality of oxy-acetylene pressure-welded joints in thick-walled, high-strength, alloy-steel tubing. With suitable machines and auxiliary equipment, joint types, and protective atmospheres and coatings, it was found possible to produce unreinforced joints with strengths equal to the parent metal, even when heat-treated to tensile strengths in the region of 150,000 lb./sq. in. Consistent results were, however, not obtained because of the presence of an oxidized zone of weakness at the weld plane.—C. O.

Oxyacetylene Pressure Welding of Some Low-Alloy Steels and Non-Ferrous Alloys. K. H. Koopman. (Welding Journal, 1946, vol. 25, Nov., pp. 1069–1076). The results of tests on welds in low-alloy steels and non-ferrous alloys are shown in tables and by photomicrographs. The welds were made by heating the welding zone to a temperature of 1200–1250° C. with a multiple-tip oxy-acetylene torch while maintaining a pressure of 2,500 lb./sq. in. between the faces being joined. Nickel–chromium–molybdenum steels, such as NE 8740 and NE 8630, manganese–molybenum steels, and silicon spring steels of high-carbon content had high tensile strengths after welding, but a quenching and tempering treatment was required to restore high impact resistance. Steel containing 5% of chromium and 0·50% of molybdenum was pressure-welded

successfully, optimum ductility in the weld zone being restored by annealing.—C. O.

The Argonarc Process for Welding Magnesium and Aluminium Alloys and Stainless Steel. R. E. Dore. (Sheet Metal Industries, 1946, vol. 23, Dec., pp. 2405–2415, 2420). A detailed description is given of the technique and equipment used in welding aluminium and magnesium alloys and stainless steels by the Argonarc process. Heat is generated by an electric arc struck between a tungsten electrode and the work, inside a protective envelope of argon gas. Filler rod, if needed, is introduced into the weld as in oxy-acetylene welding.—C. O.

Welding Stainless Steel by the Argonarc Process. W. A. Woollcott and R. R. Sillifant. (Welding, 1946, vol. 14, Dec., pp. 557–565). The equipment and technique used in welding stainless steel by the Argonarc process are described and illustrated in detail. The source of heat is an arc struck between the work and a carbon or tungsten electrode, inside a protective atmosphere of the inert gas, argon. Micrographs are shown of manual and automatic welds in stainless steel sheets.—C. O.

Arc Welding of High-Speed Steel. D. D. Howat. (Welding, 1946, vol. 14, Dec., pp. 544–556). Descriptions are given of methods successfully applied in the repair of broken high-speed cutting tools, involving pre-heating with an oxy-acetylene torch and arc-welding. The properties of butt-welds between sheets of mild and high-speed steel were experimentally investigated. It was found that very careful control of the pre- and post-heating treatments was necessary to ensure freedom from brittleness.—C. O.

Causes and Prevention of Defects in Welding. F. S. Dever. (Welding Journal, 1946, vol. 23, Sept., pp. 860–864). The defects which may occur in welds made by the arc-, gas-, atomic hydrogen-, and spot-welding processes are summarized, and the faults in technique to which each may be attributed are explained briefly. —C. O.

German Gas Welding and Cutting Industry. (Combined Intelligence Objectives Sub-Committee, 1946, File No. XXXIII–49 : H.M. Stationery Office). A detailed illustrated account is given of German gas-welding and cutting practice during the war.—C. O.

Oxygen Cutting. E. S. Semper. (Welding, 1946, vol. 14, Oct., pp. 469–473 ; Nov., pp. 520–526 ; Dec., pp. 570–574). These papers are a continuation of a series of articles on modern cutting practice (*see* Journ. I. and S. I., 1946, No. 11, p. 97A). Operating data are given for the use of machine-controlled oxy-acetylene cutters.—C. O.

Some Recent Canadian Contributions to Oxygen Cutting. R. A. Dunn. (Welding Journal, 1946, vol. 25, Oct., pp. 946–952). Illustrated examples are given of advances in oxygen-cutting practice made in the production of armaments in Canada during the war. —C. O.

Oxygen-Arc Cutting. (Steel, 1946, vol. 119, Dec. 2, pp. 116–118). A short description is given of a process for cutting metals by means of an arc struck between the metal and a coated mild steel tube down which oxygen is passed. Temperatures 3000° F. higher than those of the usual cutting processes are developed.—C. O.

Arc Welding and Cutting under Water. L. Mills. (Transactions of the Institute of Welding, 1946, vol. 9, Aug., pp. 128–130 ; Oct., pp. 156–158, 167). The equipment and technique for underwater welding are described.—R. A. R.

CLEANING AND PICKLING

Infra-Red Degreasing. (Steel, 1946, vol. 119, Nov. 18, pp. 104–139). A process for burning grease from metal tubes by infra-red heating is briefly described.—C. O.

Postwar Metal Cleaning Methods. L. De Waltoff. (Metal Progress, 1946, vol. 50, Nov., pp. 1064–1065). A very brief summary is made of the available metal cleaning methods.—C. O.

Principles of Metal Cleaning. J. C. Harris. (Metal Progress, 1946, vol. 50, Nov., pp. 1063–1064). The development of improved metal-cleaning processes is discussed.—C. O.

A Review of Inhibitors. W. G. Imhoff. (Wire and Wire Products, 1946, vol. 21, June, pp. 447–450, 478–482 ; July, pp. 520–523, 542–547 ; Aug., pp. 598–599, 621–623 ; Sept., pp. 673–674, 699–701). Other investigators' work on the use of inhibitors in pickling baths is reviewed and an extensive bibliography is presented.—R. A. R.

PROTECTIVE COATINGS

Reducing Embrittlement in Electroplating. C. T. Eakin and H. W. Lownie, jun. (Iron Age, 1946, vol. 158, Nov. 25, pp. 69–72). Investigations on the embrittlement of steel springs during electroplating are described. It was found that agitation of the electrolyte reduced embrittlement by causing the hydrogen given off to form bubbles instead of being absorbed by the metal. Complete relief from embrittlement of cadmium-plated springs is obtained by heating for more than 1 hr. at 527° F. Electroplating can be removed

by non-acid solutions such as ammonium nitrate without embrittlement of the underlying steel. —C. O.

A Periodic Chart for Electroplaters. G. Dubpernell. (American Electroplaters' Society : Iron Age, 1946, vol. 158, Nov. 28, p. 51). The use of a periodic chart designed to show the relations between elements which can be electrodeposited as metals from aqueous solutions is described.—C. O.

Wartime Electroplating. G. B. Hogaboom. (Metal Progress, 1946, vol. 50, Nov., pp. 1065–1067). A review is made of developments and applications of electroplating during the recent war.—C. O.

Protection against Corrosion by Chromium Diffusion Zone. G. Becker, K. Daeves, and F. Steinberg. (Iron and Steel Institute, 1947, Translation Series, No. 296). This is an English translation of a paper which was published in Zeitschrift des Vereines deutscher Ingenieure, 1941, vol. 85, Feb. 1, pp. 127–129 (*see* Journ. I. and S. I., 1946, No. II, p. 30A).—D. R. S.

The Production of Chromium Coatings by Diffusion. H. Bennek, W. Koch, and W. Tofaute. (Iron and Steel Institute, 1947, Translation Series, No. 300). This is an English translation of a paper which was published in Stahl und Eisen, 1944, vol. 64, Apr. 27, pp. 265–270. (*See* Journ. I. and S. I., 1945, No. I, p. 25A). —D. R. S.

Report on Investigation of Methods of Gaseous Metal Treatment. (British Intelligence Objectives Sub-Committee, 1946, Final Report No. 839 : H. M. Stationery Office). An account is given of investigations at a number of German works into the " chromizing " process. The method developed and operated on a commercial scale at the Metall und Diffusionsgesellschaft, Düsseldorf, was devised by G. Becker, F. Steinberg, and K. Daeves before the war. They described it as follows : A "chromizing mass " is prepared by generating chromium chloride and absorbing it in a ceramic material ; the chromium chloride is formed by passing hydrogen-chloride gas over a mixture of equal volumes of broken, unglazed crockery and ferrochromium containing 65% of chromium, heated together for 5 hr. at 1000° C. in an inclined retort, 10–15ft. in length and 2–3ft. in dia.; the mass, when cool, is stored for use. The articles to be chromized are degreased and packed in a similar retort with the chromizing mass. Air is swept from the retort by a stream of hydrogen, and then a temperature of 1000–1050° C. is maintained for 5 hr., during which time the reaction

$$Fe + CrCl_2 \longrightarrow FeCl_2 + Cr$$

progresses. The chromium diffuses into the steel, reaching a maximum surface concentration of 35–40%. The depth of diffusion, which is taken as the depth to which an acid-resistant phase is present, is about 0·1 mm. under standard time and temperature conditions. Temperature is the more influential factor in increasing the depth of diffusion, which is also affected by carbon content owing to the formation of chromium carbide which obstructs diffusion. Special chromizing steels have been developed, one, known as IK.1 containing 0·06% of carbon and 0·5% of titanium, and another 3% of manganese and 1% of carbon. Stainless steel (18/8) also chromizes satisfactorily. Several firms have used this process under licence, sometimes with modifications, such as the combination of the chloride generation and the treatment at the Rudolph Otto Meyer works at Hamburg.

Two other methods had been used by the Krupp concern. The first involved the immersion of the article in a bath of molten chlorides (60% of chromium chloride, 40% of barium chloride) for 4 hr. at 1100° C. A coating 0·14 mm. in thickness was produced, but crucible corrosion problems caused the process to be abandoned commercially. The second method was to heat the steel for 4–5 hr. at 1100° C. in an electrically heated retort, evacuated to a pressure of 20 mm., and containing chromium chloride in boats on the floor. The results were very similar to the Becker, Daeves, and Steinberg process.—C. O.

The Nitriding of Steel. C. F. Floe. (Metal Progress, 1946, vol. 50, Dec., pp. 1212–1220). A comprehensive account is given of the process of surface-hardening steels by nitriding. Particular reference is made to the gaseous atmospheres and the furnaces used commercially, the metallography of the nitride layer, the types of steel suitable for nitriding, and industrial applications. —C. O.

Hot-Dip Galvanizing Practice. W. H. Spowers, jun. (Steel, 1946, vol. 119, Aug. 12, pp. 108–110, 130, 131 ; Aug. 26, pp.100–106 ; Sept. 2, pp. 127–128 ; Sept. 9, pp. 136–140, 155 ; Sept. 16, pp. 122–124 ; Sept. 23, pp. 149–153 ; Sept. 30, pp. 74–78, 102–105 ; Oct. 7, pp. 134–137, 172–174 ; Oct. 21, pp. 118–121, 137–140 ; Oct. 28, pp. 114–116, 126–130 ; Nov. 4, pp. 118–120 ; Nov. 18, pp. 112–114 ; Nov. 25, pp. 88–92 ; Dec. 2, pp. 123–129). The author published a book on hot-dip galvanizing practice in 1938. This has now been re-edited, abridged, and brought up-to-date to cover the latest developments in

hot-dip galvanizing. The new edition is presented in these 14 parts.—R. A. R.

Nickel-Zinc Corronizing. L. C. Conradi. (Corrosion and Material Protection, 1946, vol. 3, Aug.-Sept., pp. 12–13). The process of nickel-zinc Corronizing is described briefly. A layer of nickel is electroplated on to the base metal (steel, brass, or bronze), followed by a layer of zinc, and the article is then heated to about 700° F. to allow diffusion of the zinc and nickel. The main application of Corronizing is for the protection of metals from attack by sulphurous gases.—C. O.

Calorizing. B. J. Sayles. (Corrosion and Material Protection, 1946, vol. 3, Aug.-Sept., pp. 11–12). A brief account is given of the applications of the process known as Calorizing, in which a corrosion-resistant alloy of aluminium and iron is formed on the surface of steel.—C. O.

Alitised Iron As Sulphur Resistant Material. J. A. Verö. (Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy, 1940, vol. 12, pp. 158–166). It was desired to find a metal for making baskets for coking briquettes in a low-temperature carbonization plant. The material had to be resistant to a reducing atmosphere containing H_2S at temperatures up to 600° C. "Alitized" steel was found to be more resistant than high-chromium steel. Steel is alitized by heating it for 3 hr. or longer at 930–950° C. embedded in a mixture of burnt alumina 49 parts, 55/45 aluminium-iron alloy 49 parts, and ammonium chloride 2 parts; hydrogen is passed slowly through the tubular furnace during the process. At room temperature the alitized layer is somewhat brittle, but at 500° C. it can be bent sharply without cracking. Data on the properties of alitized steel wire at temperatures in the 400–900° C. range are presented and compared with similar data for mild steel and chromium steel wire.—R. A. R.

Zinc Spraying. J. Howat. (Corrosion and Material Protection, 1946, vol. 3, Aug.–Sept., pp. 8–10). The advantages of the zinc-spraying process over other methods of applying zinc coatings to steel are discussed, and illustrated examples are given of its use.—C. O.

Metal Spraying—The Process and Industrial Applications. J. Barrington Stiles. (Welding, 1946, vol. 14, Nov., pp. 493–499). The principles of the metal-spraying process are described in detail, with particular reference to the preparation of the surface to be sprayed. Among the numerous applications mentioned are: (1) Protective coating; (2) the building-up of worn parts, such as crank shafts; (3) the reclamation of faulty castings; (4) decorative finishing, *e.g.* the spraying of wood, glass, and plaster; and (5) coating to give conductivity, *e.g.* the spraying of paper for electostatic condensers.—C. O.

Spray-Welding. (Steel, 1946, vol. 119, Dec. 23, pp. 74–77). A brief description is given of a method of applying coatings of hard-facing alloys. A gun has been devised in which powdered facing-alloy is blown by compressed air on to the base metal and simultaneously fused in an oxy-acetylene flame.—C. O.

Factors Governing Adherence of Enamels Applied to Sheet Iron. L. A. Johnson and E. E. Howe. (Journal of the American Ceramic Society, 1946, vol. 29, Oct. 1, pp. 296–301). An account is given of tests conducted to ascertain the effects of nickel depositions and variations in firing conditions upon the adherence of enamels to two special enamelling steels. The optimum coating of nickel was found to be 0·08–0·10g./sq. ft. in both cases. The three enamels used, zirconium-opacified white, antimony-molybdenum-opacified white, and cobalt nickel blue, showed wide variations in the degrees of adherence, which were attributed to the oxidation characteristics of the steels and the differing solubilities of iron oxide in the enamels.—C. O.

Fabrication and Enameling Performance of Titanium Enameling Steel. F. R. Porter. (American Ceramic Society Bulletin, 1946, vol. 25, July 15, pp. 259–266). The results are given of investigations made during the fabrication of enamelled articles from a basic open-hearth steel containing a maximum of 0·09% of carbon and 0·20–0·50% of titanium. The steel is non-strain-ageing and therefore does not show strain lines when drawn; it is highly resistant to hydrogen penetration, and enamels applied to it show no fish-scale cracking even if there is no adherence. It is inherently harder than enamelling iron. The principal advantage obtained by using the steel is the elimination of the necessity for a ground coat, with consequent decrease in time and labour required in the enamelling process.—C. O.

Plastic Packaging for Metal Parts. (Steel, 1946, vol. 119, July 22, p. 110). Brief details are given of a plastic called "Seal-Peel" which is used as a protective coating for metal parts. The part is dipped into the heated plastic; on withdrawal, the plastic dries in a few seconds to form a thin tough transparent coating which is stable at temperatures between −120° and +160°F.—R.A.R.

Plastic Coatings to Control Metal Corrosion. S. P. Wilson. (Electrochemical Society, American

Chemical Society, and American Institute of Chemical Engineers, Joint Symposium: Steel, 1946, vol. 119, Oct. 14, pp. 150–156). The properties and applications of plastics, both thermosetting and thermoplastic, which are suitable for coating metals, are given.—C. O.

Gel Lacquer Coatings. C. J. Malm and H. L. Smith, jun. (American Chemical Society: Steel, 1946, vol. 119, Oct. 28, pp. 77, 98, 100, 105). **Gel Lacquer Technique for Protective Coating.** C. J. Malm and H. L. Smith, jun. (Industrial and Engineering Chemistry, 1946, vol. 38, Sept., pp. 937–941). The gel-lacquer technique is designed for the application of unusually heavy coatings. The process depends upon the property of the lacquer to undergo gelation solely by a change in temperature without the necessity for the evaporation of a solvent. This behaviour is obtained by proper selection and balance of the solvents. Film formers chosen from the group of cellulose mixed esters are suitable for the base of these lacquers, and produce coatings of plastic appearance and properties. The lacquer is applied by dipping in a viscous solution maintained at approximately 120° F. On cooling to room temperature the heavy coating immediately sets to a firm gel which will not sag or wrinkle.—C. O.

Phosphate-Coated Car Bodies. (Machinery, 1946, vol. 69, Oct. 31, pp. 545–548). An illustrated description is given of the production methods used by the Ford Motor Company in rust-proofing all the body parts of vehicles by their phosphate-coating process. The preparation of surfaces by cleaning and drying with infra-red rays is of great importance and is described in detail.—C. O.

Recommendations for Phosphate Coatings as a Basis for Painting Steel. (British Standards Institution, 1946, Aug., No. PD 539).—R. A. R.

Modern Drying Practice with Particular Reference to Radiant (Infra-Red) Heating. W. G. J. Appleton. (Journal of the Institution of Production Engineers, 1946, vol. 25, Oct., pp. 265–289). The theory of infra-red or radiant heating is explained, and this form of heating is compared with other systems which have been used in industry for many years. The infra-red system is efficient and economical where parts of plain section, light gauge, and identical shape are to be stoved, particularly when a sprayed coating is to be dried, but for parts of intricate shape and of varying or heavy section, a well-designed air-circulated drying system is better. Generally speaking, a forced convection stove gives better results with dip-coated parts than a radiant type unit. Gas and electric radiant plants are compared; the author favours the former, which has higher energy densities, uses a cheaper fuel, and is more flexible from the control point of view.—R. A. R.

Production of Clad Steels by Rolling. C. H. Wick. (Machinery, 1946, vol. 69, Nov. 7, pp. 584–589). A detailed and illustrated description is given of a method of "cladding" steel plate with a sheet of nickel. The same method can be used for bonding Inconel, Monel, or stainless steel to mild steel. The process is known as the Jessop method, and it has been described previously (*see* Journ. I. and S. I., 1942, No. II., p.113A).—C.O.

Automatic Spray Enameling. (Steel, 1946, vol. 119, Oct. 7, p. 124). A brief illustrated description is given of a machine developed by the Ferro Enamel Corporation, Cleveland, Ohio, for the spraying of porcelain enamel on the inside of hot water tanks.—R. A. R.

The Influence of Cobalt and Nickel in Vitreous Enamel Ground-Coats. G. C. Priddey. (Institute of Vitreous Enamellers: Foundry Trade Journal, 1946, vol. 80, Nov. 14, pp. 263–268, 271). The principal theories which have been put forward to explain the nature of the bond between vitreous enamels and iron and steel are discussed with particular reference to the action of the oxides of cobalt and nickel. These latter are certainly instrumental in the promotion of adherence, but the mechanism of the process is not yet established.—C. O.

Influence of Chemical Composition of Enamel on a Few of Its Properties. B. K. Niklewski. (Journal of the American Ceramic Society, 1946, vol. 29, Nov., pp. 316–331). A simple method was developed for determining the fusibility of enamels with an accuracy of $\pm$ 5° C. It was used to study the effects of adding Fe_2O_3 to boron-free and boron-containing ground coats for sheet metal. Boron-containing enamels show higher solubility of the iron oxide and a smaller change in fusibility with increasing iron-oxide content, they therefore have better working properties than the boron-free ground coats. The strength of the bond between various ions and oxygen atoms was calculated according to the Kossel theory.—R. A. R.

Observations on Some Surface Properties of Vitreous Enamels. W. E. Benton. (Institute of Vitreous Enamellers: Foundry Trade Journal, 1947, vol. 81, Jan. 2, pp. 11–16). Theoretical aspects of the property of enamels known as gloss or brilliance are discussed. The refractive index and the smoothness of the surface are the two most influential factors. Mention is made

of methods of measuring the smoothness of enamel coatings on steel surfaces.—C. O.

The German Vitreous Enamel Industry. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 788 : H. M. Stationery Office). After visits to many German enamelling plants, a report is presented on the state of their frit, colour and enamelling industries.—R. A. R.

New Coatings and Applications Expand Use of Porcelain Enamel. (Steel, 1946, vol. 119, Nov. 11, pp. 154–156. 218–222). A summary is given of the domestic and industrial uses and possible future applications of porcelain enamel coatings. —C. O.

Postwar Organic Finishes. G. Klinkenstein. (Metal Progress, 1946, vol. 50, Nov., p. 1076). Wartime developments in the use of organic coatings on metals are reviewed.—C. O.

Postwar Paints and Painting Methods. E. A. Zahn. (Metal Progress, 1946, vol. 50, Nov., p. 1075). Recent developments in painting technique are briefly reviewed.—C. O.

Postwar Metal Finishing. A. Bregman. (Metal Progress, 1946, vol. 50, Nov., pp. 1061–1063). Recent trends and developments in the coating of metals with lacquers and paints are reviewed. —C. O.

Measurement of the Adhesive Force of Lacquers. P. Koole. (Sheet Metal Industries, 1946, vol. 23, Nov., pp. 2132–2134). A description is given of a simple apparatus for measuring the adhesive force of a lacquer, which is defined as the force necessary to scrape off a strip of lacquer 1 cm. wide from an underlayer by means of a chisel. —C. O.

Efficiency of Conveyor Drying and Enamelling Plants. F. J. Ballard and Co., Ltd. (Ministry of Fuel and Power, Fuel Efficiency Committee Conference on "Fuel and the Future," London, Oct., 1946). Data are given on the performance and operation of radiant heat and air-circulated direct-fired convection conveyor-type ovens.—C. O.

Static Makes Dipped Coatings Shed Tears. E. P. Miller. (Machinist, 1947, vol. 90, Jan. 4, pp. 1584–1586). An illustrated description is given of a process for removing excess paint or other liquid coating from metal parts by the attraction of electrostatic fields.—C. O.

Rubber-Metal Composites. J. A. Merrill. (Materials and Methods, 1946, vol. 24, Oct., pp. 891–895). Details are given of the properties and applications of a number of rubber-like materials and plastics, resistant to attack by petroleum, oil, and acids, and suitable for use as protective coatings for metals. Illustrated examples of their application are shown.—C. O.

POWDER METALLURGY

Powder Metallurgy from the Engineer's Point of View. J. F. C. Morden. (Coventry Engineering Society Journal, 1946, vol. 27, Nov.–Dec., pp. 92–102). The technique of powder metallurgy is reviewed briefly, and its importance to the engineer explained, with a warning as to popular misconceptions of the scope of its applications. Methods of powder production, and of the manufacture of components from metal powders, are described, and the principal applications summarized. Examples are given of forms suitable for production by powder-metallurgy methods, and the main advantages and limitations of the technique are listed. The physical properties of typical bronze and iron alloys, and running-test figures obtained with porous bronze bearings, are tabulated.—M. A. V.

Seminar on the Theory of Sintering. F. N. Rhines. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2043 : Metals Technology, 1946, vol. 13, Aug.). This paper gives a review of the literature on the sintering of pure metal powders. The information is classified according to experimental observations and theory, and presented in summary form. The theory is a composite one, developed to take into account all the established facts. An appendix contains 62 references.—C. O.

The Pressing Operation in the Fabrication of Articles by Powder Metallurgy. R. P. Seelig and J. Wulff. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2044 : Metals Technology, 1946, vol. 13, Aug.). An account is given of the theory and practice of the pressing operation in powder metallurgy. It is convenient to regard the process of pressing in three phases, *i.e.*, packing, elastic and plastic deformation, and cold-working, with or without fragmentation. There is no clear demarcation between the phases, which may be proceeding simultaneously in different parts of the same compact. In production, powders react differently according to purity, particle size, size distribution and shape, and method of manufacture. Such properties as "compactibility" and green strength are of practical importance, but are not easily defined or measured. A study of stress distribution during the pressing cycle shows that high pressures are required only very shortly before the maximum stroke is reached. The questions of

powder feeding to the die and the construction of the die are dealt with briefly.—C. O.

A Study of the Physical Properties and Microstructure of Sintered Steel. G. Stern. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2045 : Metals Technology, 1946, vol. 13, Aug.). An account is given of the properties of carbon steels made by powder metallurgy using intimate mixtures of iron powder and graphite. Electrolytic iron powder plus graphite gives steels with similar yield points and tensile strengths to S.A.E. plain carbon steels, but with lower ductility and impact resistance because of residual porosity. The properties of such sintered steels may be altered in the same way as those of ordinary steels by heat-treatment, but a dry and neutral atmosphere is necessary. Steels in $\frac{5}{8}$-in. square bars made from electrolytic iron powders having carbon contents from 0·28% to 0·87% do not transform to give martensite when oil- or water-quenched. Samples slowly cooled from the sintering temperature sometimes show coarse pearlite with an occasional isolated individual lamella because of the high purity of the steel. The superiority of the properties of steels made from electrolytic powders is explained on the basis of density values and microstructure.—C. O.

Nickel–Iron Alloys Produced by Powder Metallurgy. L. Delisle and A. Finger. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2046 : Metals Technology, 1946, vol. 13, Aug.). A description is given of the preparation and testing of specimens of nickel–iron alloys in the range 0–100% of nickel. The test bars were prepared from mixtures of nickel and electrolytic iron powders, compacted in hydraulic presses and sintered at 1100° C. The physical properties of the alloys followed the same trend as those prepared conventionally, with critical values in the range of 15–20% of nickel. The microstructures showed no pure iron to be left in specimens containing more than 70% of nickel, so that diffusion in the solid state must have proceeded to a great extent. The 20/80 nickel–iron alloy had a tensile strength of 106,000 lb./sq. in. and 4·3% elongation after two pressing and sintering operations. This is considered to reflect the $\gamma \rightarrow \alpha$ transformation in the binary system, which gives a substantial increase in strength even in the presence of only very small amounts of carbon.—C. O.

Some Considerations in Designing Parts for Powder Metallurgy. I. J. Donahue. (Mechanical Engineering, 1946, vol. 68, Nov., pp. 949–952). Illustrations are given of some of the fundamental principles of the design of parts to be produced by powder-metallurgy.—C. O.

The Uses of Powder Metallurgy in Automobile Engineering. Part 1 : Ferrous Powder Metallurgy. J. A. Judd. (Institution of Automobile Engineers, 1946). After a brief review of the techniques used in preparing iron powders and powder compacts, the limitations of powder metallurgy as a commercial process are considered. The main ones are : Lack of plasticity in metal powders, which restricts lateral flow in the die and so makes difficult the production of grooves or undercuts ; the friction between the powder and the die wall, which restricts the ratio between the length and width of the compact ; the abrasive action of the powder on the die walls, which affects the permissible clearances and tolerances ; and the lack in this country of good-quality, low-priced powders. The articles which have up to now been produced commercially include bushes, valve guides, oil-pump gears, electrical components, and piston rings for internal combustion engines. The future of ferrous powder metallurgy is considered to be largely in the hands of the designer, who should be able to design parts specifically for the process rather than adapt to it components designed for production by more conventional methods.—C. O.

PROPERTIES AND TESTS

Control of Welding Stresses by Welding Sequence. P. Ffield. (Welding Journal, 1946, vol. 25, Nov., pp. 1109–1122). A detailed discussion is given of two researches undertaken by the Shipbuilding Division of the Bethlehem Steel Company. The first was to determine the characteristic pattern of the residual stresses in a welded ship, and the second was to investigate the possibility of using these stresses to strengthen rather than weaken the structure. The measurements were made by electrical strain gauges on a 28-ft. long model of a Liberty ship and on welded plates of a destroyer under construction. A comparison of results indicates that the three-dimensional model is a reliable guide to the behaviour of its full-scale counterpart.—C. O.

Inspection and Testing for the Heat Treatment Shop. Part 2. Inspection and Testing. L. R. Flewin. (Australian Institute of Metals : Australasian Engineer, 1946, Oct. 7, pp. 82–88). A general description is given of the methods of mechanical testing and inspection for defects used in the Victorian Railway Workshops, Newport, Australia.—C. O.

Determination of Limit of Proportionality of Wires. F. C. Thompson and W. R. Tyldesley.

(Nature, 1947, vol. 159, Jan. 4, p. 30). A brief description is given of a novel method of determining the limit of proportionality of cold-drawn wire, in which the wire itself is employed as a strain gauge, the increase of electrical resistance with increasing load being measured. Load/increase-in-resistance curves obtained on mild steel and a nickel-copper alloy are reproduced, and it is shown that they closely follow the normal stress-strain curves of these materials.—D.R.S.

Causes and Prevention of Intergranular Fracture in Cast Steel. C. H. Lorig. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 234–240). A report on an investigation of intergranular fracture in cast-steel armour plate is presented. The chief causes were found to be the precipitation of aluminium nitride and of a network of ferrite at the primary grain boundaries.—R. A. R.

The Appearance of Rock Candy Fracture. M. F. Hawkes. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1945, vol. 3, pp. 240–244). Investigations of fractures in low-alloy cast steel which support the conclusion of C. H. Lorig (*see* preceding abstract) are reported.—R. A. R.

Symposium on the Hardenability of Steel. (Iron and Steel Institute, 1946, Special Report No. 36). This Symposium was organized by the Technical Advisory Committee of the Special and Alloy Steels Committee (Ministry of Supply) and the Iron and Steel Institute. It gives an account of the work done by the Hardenability Sub-Committee of the Technical Advisory Committee which was set up on the return of a small mission of British metallurgists from their discussions in the United States and Canada on means of conserving critical alloys used in steel production. This Sub-Committee had three principal tasks, namely : (*a*) To standardize the Jominy end-quench hardenability test ; (*b*) to examine the fundamental principles governing the test and recommend the best methods of interpreting the results ; and (*c*) To survey a representative set of B.S. En steels by hardenability test, and correlate the results with the other mechanical properties.

The Symposium, comprising 430 pages, is in ten Sections. The first is introductory in character. Section II is in two parts, dealing respectively with the effect on Jominy hardenability curves of changes in some of the test conditions, and the reproducibility of these curves. Section III gives full details of the standard procedure for carrying out the Jominy end-quench test. Section IV is in three parts. In the first the theoretical aspects of the end-quench tests are considered ; the second reports an investigation of the surface temperature in the cooling of a Jominy test-piece ; and in the third Jominy test results are correlated with data from quenched cylindrical bars. Section V is in five parts. These are concerned with : (*a*) The significance of variation in the Jominy hardenability of some hardened and tempered En steels ; (*b*) the influence of variations in composition on end-quench hardenability ; (*c*) the variation in end-quench hardenability between casts of each of several En steels ; (*d*) the variation in end-quench hardenability of different samples from the same cast of steel ; and (*e*) the effect of the location of the test-piece on the Jominy hardenability of billets. Section VI, also in five parts, is devoted to investigations of the relationship between the Jominy hardenability and the mechanical properties of several steels representing the different types in the En series. Studies of (*a*) the effect of the major alloying elements and of boron on the hardenability, and (*b*) the hardenability of a hypereutectoid chromium–molybdenum steel are reported in Section VII. A summarized statement of the information to be derived from the end-quench hardenability curve together with the principal conclusions resulting from the Sub-Committee's work are given in Section VIII. An account of the Sub-Committee's activities and suggestions for future research are set forth in Section IX. Finally, Section X consists of a bibliography of 255 references covering the literature from 1938 to 1944. Abstracts of the papers follow.—R. A. R.

The Effect on Jominy Hardenability Curves of Varying Certain of the Test Conditions. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 9–17). A series of investigations was carried out to determine the degree of latitude permissible in several of the experimental details of the Jominy end-quench tests. The results indicate that the hardenability curve is not appreciably affected by : (1) Varying the temperature of the quenching water over the range 4–60° C. (2) Varying the "free" height of the jet from 1 to 20 in. (3) Varying the diameter of the jet orifice from $\frac{1}{8}$ to $\frac{5}{8}$ in. (4) Varying the distance of the test-specimen above the jet orifice from $\frac{1}{4}$ to $\frac{3}{4}$ in. (5) Decreasing the length of the test-specimen from 4 to 3 in. (6) Quenching the entire test-specimen after end-quenching for only 10 min.

Reproducibility of Jominy End-Quench Hardenability Curves. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 17–23). End-quench hardenability curves of three steels, determined in different laboratories have been compared. The maximum variation noted was of the order of $\pm 0{\cdot}07$ in. at that part of the curve where the hardness changes rapidly. Further tests indicated that the error introduced during grinding and hardness

testing by different observers and laboratories was negligible.

Standard Jominy End-Quench Test Procedure. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 23-25). Full details are given of the standard procedure for carrying out the Jominy end-quench test.

Some Theoretical Considerations of the Jominy Hardenability Test. T. F. Russell. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 25–33). This paper is in two parts. In the first, a study of the theoretical methods used to convert the results of a Jominy end-quench test to critical diameters leads to the conclusion that too much depends on how the rate of cooling is defined. Three cases are considered: (1) Rate of cooling at some particular temperature, (2) time to reach some particular temperature from the quenching temperature, and (3) average cooling rate over some particular range of temperature. For each criterion of cooling rate the calculated critical diameter for any intensity of quench varies appreciably with the particular temperature or range of temperature chosen. The best choice may depend to a large extent on the type of steel under consideration.

In the second, numerical data are given which simplify the investigation of correlation between the Jominy end-quench test and the hardening of round bars, based on the half-temperature time as a criterion of cooling rate.

Surface Temperature Measurements during the Cooling of a Jominy Test-Piece. T. F. Russell and J. C. Williamson. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 34–46). Temperature measurements have been made during Jominy end-quench tests. The effects on the cooling curve of position along the test-piece, composition of the steel used, and initial temperature have been studied. From the measurements of half-temperature times, values of the thermal diffusivity and the heat-transfer coefficients have been deduced. A graphical method, based on equal half-temperature times, has been devised for the easy approximate solution of various problems relating to the correlation between Jominy tests and the quenching of round bars.

Correlation between Jominy Test Results and Quenched Cylindrical Bars. W. I. Pumphrey and F. W. Jones. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 47–56). The transformation range has been determined for a number of hypereutectoid steels during continuous cooling at such rates as are attained at points along a Jominy bar. The application of the results of these determinations to the correlation of Jominy curves with the hardness obtained on quenched round bars is discussed.

The Significance of Variation in Jominy Hardenability with Respect to the Mechanical Properties of Some Hardened and Tempered B.S. En Steels. H. Allsop. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 56–81). Several B.S. En steels have been examined with respect to hardenability expressed as critical diameters, calculated from composition and grain-size and deduced from end-quench curves. Comparison of the values obtained by the different methods does not show very close agreement.

The correlation between the end-quench hardenability and the mechanical properties of the heat-treated steels has been examined. Appreciable variation in end-quench hardenability has been observed between the steels of each of the B.S. En types considered, but the variation appears to have little significant effect on the mechanical properties of the steels heat-treated to the specification requirements. The evidence indicates that tempering at temperatures such as those normally used for the steels examined largely nullifies the hardenability variation shown by the steels of each type.

The relationship between as-quenched hardness and Izod impact value of tempered bars of a number of B.S. En 17 steels is discussed, and the possible influence of inherent cast properties, other than hardenability, is mentioned.

The limited data presented reveal a marked discrepancy between the actual internal hardness of oil-quenched bars and that derived from end-quench curves using a widely quoted end-quench-curve/bar-size relationship.

The Influence of Variations in Composition of the Order of those Permitted by Specifications on End-Quench Hardenability of Steels. W. E. Bardgett. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 81–99). Results are recorded to show the influence of variations in composition, of the order of those permitted by specifications, on the end-quench hardenability of a number of selected qualities of steel. Comparisons of the ideal critical diameter calculated from chemical composition and derived from the end-quench hardenability curve have been made. The agreement is in general fairly good.

The Variation in Hardenability between a Number of Casts of Steel to Each of Several B.S. En Specifications. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 99–113). In order to obtain some indication of the degree of variation in end-quench hardenability of various casts of steel produced to the same specification, a number of curves were collected for each of eleven different B.S. En steels. Groups of end-quench curves are presented for B.S. En 12, 14, 16, 17, 19, 24, 25, 26, 45, 100, and 110.

A Note on the Variation of End-Quench Hardenability of Different Samples from the Same Cast of Steel. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 113–120). End-quench curves have been determined on bars of $1\frac{1}{8}$-in. dia. derived from different parts of thirteen ingots representative of five different B.S. En steels. The maximum variation in end-quench hardenability noted within the product of a single ingot was, in terms of distance from the quenched end to the steep part of the curve, an increase from 0·95 to 1·40 in. from the bottom to the top of a 12-cwt. ingot of B.S. En 22 steel. The variation in hardenability in the majority of the other ingots was considerably less, and in six ingots did not exceed the discrepancy frequently observed between the opposite flats of a single end-quench test-specimen. The variation in end-quench hardenability noted between samples of $1\frac{1}{8}$-in. dia. bar produced from different ingots of the same cast was of a similar order to that noted within the product of a single ingot.

A Note on the Effect of the Location of the Test-Piece on the Jominy Hardenability of Billets. E. H. Bucknall. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 120–131). The use of the Jominy test to decide the maximum section in which a given steel can satisfactorily be hardened is open to criticism where variations in hardenability occur within the steel, unless special precautions, not called for in current specifications, are taken to secure representative test-pieces. Examples are given of variations in hardenability across transverse sections, and the possibility of longitudinal variations

is discussed. Attention is drawn to an instance where such variations cannot be explained by local differences of composition. Finally, a simple means is described by which an adequate picture of hardenability can be secured when only transverse variations are present.

Hardenability and Mechanical Properties of a Series of B.S. En Steels. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 131–199). End-quench hardenability curves and hardness traverse curves are presented for each of a series of fifteen B.S. En steels, together with the results of mechanical tests on samples of the same steels in the oil-quenched and tempered conditions. The data appear to justify the following conclusions for steels of the types examined: (1) The hardness traverse curve of an oil-quenched bar cannot be forecast consistently with satisfactory accuracy from a single end-quench hardenability curve. (2) The hardness of an oil-quenched sample tempered in the range 500–650° C. (932–1202° F.) can be derived with satisfactory accuracy from the hardness curve determined on an end-quenched and similarly tempered hardenability test-specimen. (3) In general a reduction of hardness in the oil-quenched condition is associated with some deterioration of mechanical properties in the oil-quenched and tempered condition. The magnitude of the deterioration varies with the different types of steel, and the minimum as-quenched hardness required to develop satisfactory mechanical properties can be determined only by direct testing.

A Study of the Relationship between End-Quench Hardenability Curves and the Hardness of Bars Quenched in Oil. H. Allsop and W. Steven. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 199–252). The relationship between the end-quench hardenability curves and the hardness traverse curves of fourteen different B.S. En steels and also of seven steels of the same B.S. En type is considered. The transverse and longitudinal variations of hardenability within bars produced from three ingots of the same cast of steel are studied. The results indicate that the hardness traverse curves of different-sized bars of a group of steels of different B.S. En types, or of a group of steels of the same B.S. En type, cannot be forecast with satisfactory accuracy from end-quench curves determined on test-specimens machined from samples forged or rolled to approximately the diameter of a standard end-quench test-specimen. The main reason for the discrepancy between calculated and experimentally determined hardness traverse curves is shown to be the longitudinal and transverse variations in hardenability which exist within normal commercially homogenous steel bars. It is concluded that no single end-quench curve can be accepted as a satisfactory evaluation of the response of a given cast of steel to hardening. The hardenability of a cast of steel can be represented adequately only by a hardenability band based on a series of tests.

On the Relation between the As-Quenched Hardness and the Mechanical Properties of Quenched and Tempered Steel. A. J. K. Honeyman. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 253–267). The effect of variation in the "as-quenched" hardness on the mechanical properties of some steels in the quenched and tempered condition has been investigated. The results indicate that the notched impact strength at subzero temperatures is more susceptible than any other mechanical property and that the susceptibility increases with increasing carbon content. The Jominy end-quench test gives no guide to the minimum "as-quenched" hardness necessary to obtain satisfactory mechanical properties from a quenched and tempered steel.

Observations on the Relationship between Hardenability and the Mechanical Properties of Quenched and Tempered Steels. W. E. Cooper and N. P. Allen. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 267–283). An attempt has been made to correlate the Jominy curves and transformation characteristics of a small selection of En steels, comprising En 16, 17, 24, 25, and 26, with their ability to meet the mechanical-property requirements of typical aircraft specifications. The tests demonstrate that "hardenability" is not the only factor in the mechanical behaviour of these steels, and they suggest that other factors become increasingly important as the tensile strength required of the steels is increased.

Jominy Hardenability Tests of Low-Alloy British Standard Engineering Steels. W. Crafts and J. L. Lamont. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 283–356). Jominy hardenability tests were made on a series of B.S. En steels and the hardenability of case-hardening and heat-treating steels was found to be proportional to the recommended tensile strengths. The tensile and impact strengths of the heat-treating types of B.S. En steels were estimated from tempered Jominy specimens, and these properties were found to be quite consistent with those specified by the T.A.C. schedule (B.S. 971 (July 1941)). The validity of the correlation between the properties of tempered Jominy specimens and quenched and tempered bars up to 4 in. in dia. was established by tests on another series of alloy steels. A method for calculating the tensile strength of bars was also developed. The maximum combination of tensile strength and impact strength was found to depend on composition and grain-size, and in a given steel the maximum combination was developed only in sections that were fully martensitic after quenching.

The Effect of the Major Alloying Elements and of Boron on the Hardenability of Steel. J. Glen. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 356–400). The effect has been examined of the major alloying elements, and of boron, on the hardenability of steel, and it has been shown that no simple method of calculating the hardenability can be derived because the effect of any element is influenced by the presence of others. Graphical methods were used to illustrate the effect of the elements, both singly and in combination, on the hardenability. The ternary-alloy type of steel, especially when treated with boron, is the most economical steel to use for maximum hardenability.

The Hardenability of a Hypereutectoid Chromium-Molybdenum Steel. W. I. Pumphrey and F. W. Jones. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 400–413). The effect of quenching temperature, soaking time, initial condition, and repetition of the quenching procedure on the hardenability of a hypereutectoid chromium–molybdenum steel has been investigated. The results suggest that the amount and condition of the carbide remaining out of solution at the moment of quenching have a considerable influence on the hardenability of such a steel.

Principal Conclusions of the Hardenability Sub-Committee. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 413–416). After a study of all of the investigations the Hardenability Sub-Committee came to the following conclusions: (1) The end-quench test

is capable of yielding satisfactorily reproducible results provided that the material tested is homogeneous. (2) The results of an end-quench test are not appreciably affected by small changes in the testing conditions. (3) End-quench curves determined on different samples of bar, forged or rolled from the same ingot to approximately the diameter of an end-quench test-specimen, generally show only small differences. (4) End-quench curves determined on different samples of bar, forged or rolled from different ingots of the same cast to approximately the diameter of an end-quench test-specimen generally show only small differences. (5) A marked variation of hardenability may exist across the section of a bar. (6) The end-quench hardenability of a steel may be slightly affected by variation of the structure before quenching. (7) A wide variation in end-quench hardenability may exist between different casts of steel produced to the same specification. (8) The hardenability of a steel cannot always be calculated with satisfactory accuracy from a knowledge of the chemical composition and grain-size. (9) The hardness traverse curves of an oil-quenched bar cannot be derived consistently with adequate accuracy from a single end-quench curve. (10) The hardness of a quenched and tempered bar may be derived with satisfactory accuracy from an end-quenched and tempered test-specimen. (11) The combination of mechanical properties obtained from steel that has been quenched and subsequently tempered in the range 500–650° C. depends on the hardness developed by quenching, but there are serious difficulties in the way of interpreting end-quench curves in terms of the ability of steels to attain a required combination of mechanical properties. (12) The transformation-temperature ranges of steels during continuous cooling can be determined approximately from a series of interrupted end-quench tests.

Résumé of the Activities of the Hardenability Sub-Committee and Suggestions for Future Research. D. A. Oliver. (Iron and Steel Institute, 1946, Special Report No. 36, pp. 417–420).

End-Quench Hardenability *versus* Hardness of Quenched Rounds. G. K. Manning. (Metal Progress, 1946, vol. 50, Oct., pp. 647–652). Experiments are described in which it was sought to determine the relationship between end-quench hardenability and the actual hardness obtained in steel rounds by cooling at a given rate. End-quench tests were performed on bars cut from various parts of rounds of four low-alloy open-hearth steels of widely differing hardenability; hardness test were carried out on the remaining portions of the rounds after heat-treatment. Discrepancies were found between actual and predicted hardness of the rounds, and these tended to increase with the diameter. The reason for the lack of correlation is considered to be the presence of stresses, caused by allotropic transformations and thermal gradients during quenching.—C. O.

Boron in Steel. R. S. Archer. (Metal Progress, 1946, vol. 50, Oct., pp. 677–686). A detailed review is given of the history of the use of boron in steel up to the year 1942, and of the information then published on the effects of boron on the structure and mechanical properties of steel. —C. O.

A Note on the Influence of Nitrogen in Steel. (British Intelligence Objectives Sub-Committee, 1946, Final Report No. 898: H.M. Stationery Office). Information obtained at the Rheinmetall Borsig plant on the effect of nitrogen in steel is given. Nitrogen has been deliberately added to some steels to improve the hardenability and tempering properties. In a 12–15%-chromium steel containing 0·50% of nickel, this nickel can be replaced by 0·2% of nitrogen with an improvement in the weldability and the resistance to corrosion at the grain boundaries.—R. A. R.

The Effects of Alloying Elements on the Tensile Properties of 25-20 Weld Metal. H. C. Campbell and R. D. Thomas, jun. (Welding Journal, 1946, vol. 25, Nov., pp. 760-s–768-s). An account is given of an investigation of the effects of individual alloying elements upon the strength of stainless-steel electrode wires containing 25% of chromium and 20% of nickel. The conclusions reached were: (1) Carbon additions up to 0·20% caused an increase in tensile strength of the weld metal without serious loss of ductility; (2) sulphur, silicon, and phosphorus additions resulted in the appearance of fissures which reduced the ductility of the weld metal; (3) nickel and manganese had little effect on the tensile properties; (4) additions of the usual amounts of molybdenum, columbium, and chromium had little effect, but excessive amounts resulted in the structure becoming partially ferritic, increasing the strength, and reducing the ductility.—C. O.

The Suitability of Aluminium–Magnesium–Silicon Alloys for Killing Steel. H. Schmitz. (Stahl und Eisen, 1947, vol. 67, Jan. 2, p. 26). The results are given of impact, tensile, and stress-corrosion tests on specimens from 7-kg. melts of low-carbon steel which had been killed with aluminium–magnesium–silicon alloys containing from 11% to 40% of magnesium. All the results were satisfactory. The percentages of added and residual aluminium are given. (*Members can borrow a translation of this article from the Joint Library*).—R. A. R.

The Effect of Periods of Rest on the Fatigue Strength of Welded Joints. (Welding Journal, 1946, vol. 25, Sept., pp. 518-s–521-s). This paper is a part of the First Progress Report of the Welding Research Committee of the Engineering

Foundation of Illinois University, published as Engineering Experimental Station Bulletin, Series No. 327, in 1941. (*See* Journ. I. and S. I. 1941, No. II, p. 135A). A short account is given of the effects of rests during the fatigue testing of butt-welds in low-carbon steel plates. It was concluded that the rest periods had no significant effect upon the fatigue life of the welds when the latter were made and tested under conditions similar to those met with in ordinary constructional work.—C. O.

Quantitative Measurements of the Cracking Tendency in Welds. R. D. Stout, S. S. Tor, L. J. McGeady, and G. E. Doan. (Welding Journal, 1946, vol. 25, Sept., pp. 522-s–531-s). A description is given of the development of a test for the quantitative evaluation of the tendency of a weld to crack under given cooling conditions. The test specimen, in the form of a grooved sheet, is designed to permit variations of the restraint against which the weld shrinks during cooling. It is suggested that the degree of weld restraint may be expressed quantitatively as the " percentage of free shrinkage", *i.e.*, the ratio of the transverse shrinkage of the weld in the joint in question to the shrinkage of the weld when completely free from restraint.—C. O.

The Influence of the Melting Technique and Heat Treatment on Cracking and Resistance of Welded Sheet Steel Containing about 1 Per Cent Mn. W. Eilender, H. Arend, and I. Mircea. (Welding Journal, 1946, vol. 25, Sept., pp. 531-s–533-s). This is an abridged English version of a paper which appeared in Stahl und Eisen, 1943, vol. 63, Oct. 28, pp. 784–787 (*see* Journ. I. and S. I., 1945, No. II, p. 189A).

Variation in Quality of Spot Welds in Low-Carbon Steel over a Range of Welding Variables. J. J. Riley. (Welding Journal, 1946, vol. 25, Sept., pp. 833–842). The results are presented of a comparison of the shear strengths and macrostructures of spot welds made in low-carbon hot-rolled steel under varying conditions of welding time, electrode force, and welding current. Three combinations of the variables were used: (1) Short welding time, high electrode force, and high welding current; (2) medium welding time, electrode force, and current; and (3) long welding time, low electrode force, and low current. The results were very consistent, the shear strength decreasing and porosity increasing with increase in welding time. Characteristics such as surface indentation and distortion between welds remained approximately constant.—C. O.

METALLOGRAPHY

The Role of X-Ray Diffraction in Industrial Research. D. Goodman. (Industrial Radiography and Non-Destructive Testing, 1946, vol. 5, No. 2, pp. 32–33, 37). The position of X-ray diffraction in relation to other methods of analysis is discussed briefly. Applications of X-ray technique to the determination of structures and chemical compositions are mentioned.—C. O.

Invisible Industrial Hazard. S. R. Warren, jun. (Electrical Engineering, 1946, vol. 65, Nov., pp. 499–507). The delayed action of X-rays and gamma rays as destroyers of living tissue make these useful radiations one of the most insidious of industrial hazards. Perfect protection is not possible, but well-designed and properly installed equipment can be made safe if supplementary rules are specified and obeyed by all workers. Suggestions for such rules are given.—R. A. R.

Graphite Formation in Cast Irons and in Nickel–Carbon and Cobalt–Carbon Alloys. H. Morrogh and W. J. Williams. (Journal of the Iron and Steel Institute, 1947, vol. 155, Mar., pp. 321–371). An attempt has been made to give a concise statement of as many as possible of the features of graphite formation in cast irons which need explanation, together with a critical examination of the numerous theories which have been developed from time to time. A detailed description of the metallography of graphite in relation to its crystal structure is given, and the solidification of flake-graphite-containing irons is described.

The process of undercooling in cast irons is examined with reference to a number of problematical points and several new phenomena with respect to graphite formation in cast irons have been introduced. The effects of titanium and zirconium in refining the graphite depend upon the presence of the carbides of these elements. Two types of inverse chill have been demonstrated. The process of graphite formation in nickel–carbon, nickel–iron–carbon, and cobalt–carbon alloys has been studied, and the analogies between the mechanism of the process in these alloys and that in cast irons have been worked out. Undercooling in all these alloys and in cast irons takes place by the solidification of a eutectic of solid solution + carbide, and the carbide phase decomposes after solidification. Coarse graphite structures solidify as a eutectic of solid solution + graphite. Nickel–carbon and cobalt–carbon alloys tend to give coarse graphite when they contain more than a certain amount of sulphur. In the absence of sulphur additions, spherulitic nodular graphite structures

may be produced in nickel–carbon, nickel-rich nickel–iron–carbon, and cobalt–carbon alloys by the addition of calcium and/or magnesium in a wide variety of forms. The manner in which these spherulites form has been studied. The formation of spherulitic nodular graphite in cast irons without heat-treatment has been shown to occur. The conditions necessary for this have been outlined, and spherulites have been produced artificially without heat-treatment. The undercooling of cast irons and of similar alloys has been discussed at length, and the more general aspects of the particular type of undercooling have been emphasized.

The Microstructure of Soft Steel Deformed at High Temperature. A. Hultgren. (Northern Engineers' Meeting : Teknisk Tidskrift, 1946, vol. 76, Dec. 14, pp. 1295–1298). (In Swedish). An investigation of the red-shortness of three steels containing 0·02%, 0·004%, and 0·48% of carbon respectively, is described. Specimens 5 mm. in dia. × 25 mm. long were bent through 90° at different speeds, in the 825–875° C. range, the bending times being 0·5, 4, and 13 sec. Longitudinal sections were etched and examined under the microscope. The deformation of the two low-carbon steels caused the ferrite grains to split up into smaller units of more irregular shape with boundaries which were not flat ; etching revealed these as a network of lines or veins ; the slower the deformation the more irregular was this network. Rapid quenching after hot deformation preserved the network as sunken lines, but, after cooling slowly in air, the lines appeared to be raised. The structure of the high-carbon steel, deformed at 850° C. or higher, and cooled in air, consisted of fine lamellar pearlite and ferrite. The latter often appeared as an incomplete network indicating what had been the grain boundaries of the austenite.—R. A. R.

The Physico-Chemical Problem of the State of Metallic Surfaces. P. Bastien. (Mémoires de la Société des Ingénieurs Civils de France, 1945, vol. 98, June–July–Aug., pp. 344–356). The problem of the state of the true surface of a metal, *i.e.*, the metal/air discontinuity and immediately subjacent metal, has two essential aspects : the geometric and the physico-chemical. The latter is considered from three points of view : (1) The theoretical physico-chemical state ; (2) the real state, very different from (1) ; and (3) the principles of the methods of study of the constitution and structural state of the metallic surface. The first concerns crystal structure and growth and the intentional alteration of the surface by cementation, heat-treatment, electrolytic deposition, and chemical deposition. The second concerns the alteration in the form of the metal by solidification in a mould, by plastic deformation, and by machining, and its effects on the crystal surface. The methods of studying the state of the surface considered involve the use of the metallographic microscope, the electron microscope, X-ray diffraction, and the electronic diffraction.—*Abs. in Journ. Inst. Metals.*

Changes in Steel during Heating. V. N. Svetchnikoff. (Communications of the Kiev Polytechnic Institute, 1945, vol. 4, pp. 49–53). (In Russian). The effect of rate of heating on the position of the Ac_1 change point was studied, using high-frequency furnaces. A maximum shift of 45° C. occurred with a 0·9% carbon steel which had a laminated pearlitic structure, whilst it reached 80–85° C. with a grainy pearlite heated at 1900–2200° C./min. Generally the shift increased up to a critical rate of 200–250° C. beyond which it remained constant. The rate of heating does not seem to have any effect upon the grain size, but the physical properties are affected and, in particular, the limit of proportionality is noticeably reduced.

The solution of carbon in steels at approximately 700° C. was measured by metallography, X-ray analysis, and tensile tests and found to be 0·14–0·16% for steels specially prepared from electrolytic iron and 0·015–0·51% of carbon, and 0·08–0·10% for industrial steels with 0·08–0·95% of carbon, whilst a precise X-ray analysis at high temperatures gave values of 0·07% and 0·06% respectively. The rate of quenching has no effect upon the amount of carbon retained in solution, which is determined largely by the ratio of undissolved to precipitated carbide. Differences between the high-temperature analysis and the data obtained on the quenched specimens are ascribed to the irregular distribution of the carbon in the interior and on the surface of the grains. Even with 1·25% of manganese, 2·5% of chromium, and 0·6% of copper, the carbon could not be completely retained in solution.

Straight carbon steels and alloy steels containing 0·5%–2·5% of chromium, 1·0–2·0% of manganese, 1·5% of nickel, and 0·5–0·6% of copper were heated for from 30 sec. to 50–100 hr., the dissolved-carbon/time curves being compared with curves obtained from the tensile tests. Pearlitic cementite dissolves in austenite at a much lower rate than is generally accepted, for it was still present for 1–2 min. in a small specimen that had been rapidly heated to 800° C. Similar results were found with the alloy steels.

In 1944–45 the Kiev Polytechnic Institute was evacuated to Magnitogorsk where changes in the

sub-critical range were studied as well as the effect of operating procedure on the mechanical properties of steel. The basis of all this work is the observation of a boundary phase on the grains which occurs as a fine lattice.—R. A. H.

CORROSION

Selenium Rectifiers for Cathodic Protection. W. F. Bonner. (Corrosion, 1946, vol. 2, Nov., pp. 249–260). A detailed description is given of the construction and use of selenium rectifiers. These are suitable for cathodic-protection apparatus because of their high electrical efficiency, their adaptability to widely varying voltage and current requirements, and their ability to operate for long periods without attention and to furnish direct current with little interference from changes in temperature.—C. O.

"Thermogalvanic" Corrosion. N. E. Berry. (Corrosion, 1946, vol. 2, Nov., pp. 261–267). A type of corrosion is discussed which is caused by differences in temperature in systems which are normally stable, *e.g.*, metallic copper in contact with an acidified solution of a cupric salt and in the absence of oxygen. The term "thermogalvanic" is applied to this kind of electrochemical corrosion. A cell has been developed to enable corrosion potentials of this nature to be measured. The solubility of ferric and ferrous hydroxides limits thermogalvanic action in neutral or alkaline solutions, but it may be a prominent factor in corrosion in acid solutions.—C. O.

Maintenance of Oil Field Equipment. D. R. Hiskey. (Corrosion, 1946, vol. 2, Nov., pp. 235–247). The problem of corrosion prevention in oil-field equipment is discussed. Unless special alloys are used, it is necessary to provide protection against atmospheric corrosion for all metallic surfaces. The function of paints and plastic coatings in preventing corrosion is examined, with illustrations of actual problems encountered by the author.—C. O.

Some Metallurgical Observations with Respect to Corrosion in Distillate Wells. M. E. Holmberg. (Corrosion, 1946, vol. 2, Nov., pp. 278–285). Examples are shown of types of corrosion occurring in distillate wells. "Ringworm" corrosion is experienced in the ends of tubes which have been upset, and contain a band of fine-grained steel. It can be eliminated by a normalizing treatment.—C. O.

Bibliography of Corrosion Testing Methods, 1935–1945. L. R. Voigt. (Corrosion, 1946, vol. 2, Nov., pp. 16–29). An extensive bibliography is presented of the literature of the United States and other countries on methods of corrosion testing, with brief abstracts.—C. O.

Tin-Zinc Coatings. (Tin and Its Uses, 1946, Oct., pp. 5, 14). The results of salt-spray and humidity tests on mild steel panels coated with tin, zinc, tin–zinc alloys, or cadmium are presented and compared. The specimens coated with tin–zinc alloys, especially the high-tin alloys, offered the greatest resistance to corrosion in nearly all cases.—C. O.

Protective Value of Industrial Lubricants. G. H. von Fuchs. (American Society of Lubrication Engineers: Iron Age, 1946, vol. 158, Oct. 3, pp. 46–49). The function of lubricating oils in the prevention of corrosion and wear of metals is discussed. Rusting is usually caused by the adherence to metallic surfaces of drops or films of water. Compounds of the "surface active" type which orient themselves at the oil/metal interface prevent the adherence of water, and are therefore an essential constituent of lubricants which prevent corrosion. Illustrated examples are given of the action of rust and the use of lubricants designed to reduce corrosion to a minimum.—C. O.

Glossary of Terms Used in Corrosion. (Electrochemical Society, 1946, Oct., Preprint No. 90–12). —R. A. R.

HISTORICAL

A Short Historical Survey of the Chain Industry. H. Woodhouse. (Stourbridge Rotary Club: Wire Industry, 1946, vol. 13, Dec., pp. 673–674). After quoting mythology and biblical references to chain making the author refers to an English patent taken out by Phillip White in 1634 for making iron chains. It was not until early in the nineteenth century that the art of chain making was on a practical foundation in this country. The electric welding of chain links began 43 years ago on the Continent and has steadily developed. Medium and large stud-link cables are now made by a combination of drop forging and electric welding.—R. A. R.

The History of Coke Making. R. A. Mott. (Sheffield Coke Oven and Chemical Trades Technical Society: Coke and Smokeless-Fuel Age, 1946, vol. 8, Dec., pp. 265–266, 270). The history of the British coke-making industry up to the introduction of the modern by-product oven is reviewed.—C. O.

John Wilkinson and Bersham. R. Hughes. (Foundry Trade Journal, 1946, vol. 80, Dec. 12, p. 381). A brief account is given of the life of John Wilkinson who was mainly responsible

for the development of the Bersham Foundry, Wrexham, during the eighteenth century.—C. O.

Housing Conditions of Ironworkers in the Sixteenth Century. H. R. Schubert. (Journal of the Iron and Steel Institute, 1947, vol. 155, Mar., pp. 371–372). Leases and other old documents relating to the iron industry of the sixteenth century have been searched for information on the housing conditions of the workers of that time. The evidence indicates that only very tiny cottages of a temporary nature, sometimes roofed with earth and turf, were erected—R. A. R.

ECONOMICS AND STATISTICS

The Present Position and the Direction of the German Iron Industry. K. P. Harten and H. H. Vosgerau. (Stahl und Eisen, 1947, vol. 67, Jan. 2, pp. 2–10). An account is given of the position of the German iron and steel industry in 1946 and of the control organization which was set up under the Occupation Authorities. Details are also given of the change-over from private enterprise to state control, with data on the production of pig iron, steel ingots and rolled products. The problems confronting the industry are discussed. (*see Iron and Steel Institute, 1947, Translation Series, No. 303*).—R. A. R.

The General Position of the French Iron and Steel Industry. M. Crancée. (Mémoires de la Société des Ingénieurs Civils de France, 1945, vol. 98, June–July–Aug., pp. 279–291). This report on the French iron and steel industry covers: the sources of the raw materials, the qualities and quantities of steel which ought to be produced, existing steelworks plant and its development, distribution and prices of the products, new works and the capital required, and personnel.—R. A. R.

Review of Recent Progress in the Iron and Steel Industry of Great Britain. D. F. Campbell. (Mémoires de la Société des Ingénieurs Civils de France, 1945, vol. 98, June–July–Aug., pp. 329–343).

Problems in the Eire Iron and Steel Industry. (Iron and Coal Trades Review, 1946, vol. 153, Dec. 13, pp. 1094–1096). The information in this article was obtained from the Report of the Emergency Scientific Research Bureau set up by the Government of Eire in 1941. Brief accounts are given of techniques for the spectrographic analysis of steel, the production of ferrosilicon, the electrodeposition of iron, and the carburizing of knife blades made from mild steel. Experiments were also made with refractory materials found in Eire, and with the use of turf charcoal in foundry cupolas.—C. O.

Statistical Information. (Jernkontorets Annaler, 1946, vol. 130, No. 10, pp. 599–604). (In Swedish). Tables are presented on the production in Sweden and the Swedish exports and imports of iron ore, pig-iron, steel, and finished iron and steel goods. In most cases the figures cover the period 1937 to 1945 and the first seven months of 1946.—R. A. R.

Latin America—The Possibilities of Industrialisation. (British Iron and Steel Federation, Monthly Statistical Bulletin, 1946, vol. 21, Nov., pp. 1–9). A brief survey is given in the first part of the article of the mineral resources of Brazil. The large reserves of iron ore include deposits unsurpassed in quality by any in the world. Steel plants developed and controlled by European and American capital and enterprise have been operating in Brazil since 1860. The Brazilian Government has recently taken a leading part in the construction of the largest iron and steel works in Latin America, at Volta Redonda, which is expected to meet 60–70% of Brazil's steel requirements with its output of 270,000 tons of ingots per annum, and its very modern rail, universal plate, and continuous hot and cold wide strip mills.—C. O.

BOOK REVIEW

JENKINS, I. "*Controlled Atmospheres for the Heat Treatment of Metals.*" With a foreword by C. J. Smithells. 8vo. pp. x + 532. Illustrated. London, 1946. Chapman and Hall Ltd. (Price 50*s.*)

The improvement in the mechanical properties of alloys, particularly of ferrous alloys, by heat-treatment has vastly extended the range of usefulness of those alloys, but full commercial application of those treatments can never be made unless control is exercised over the simultaneous chemical reactions which take place on the surface of the alloy during treatment. Mr. Jenkins has provided a comprehensive guide to the theoretical and practical aspects of the control of the composition of atmospheres in contact with alloys at elevated temperatures, not only for the prevention of undesirable surface changes such as scaling and blistering, but also for the control of such processes as surface-hardening by carburizing and nitriding.

There are three distinct sections dealing respectively with the generation, the purification, and the

application of controlled atmospheres—an arrangement which should be of assistance not only to the student but also to the practical man concerned with the commercial utilization of controlled atmospheres. Each section is sub-divided into about six chapters, each of which has its own list of references appended. Further features of value to student and practical man are the many tables and illustrations, particularly of industrial plant and equipment available for the control of furnace atmospheres, and the careful consideration which has been given to the economics of the commercial maintenance of controlled atmospheres. Costs are quoted in terms of price per thousand cubic feet of gas.

The section dealing with the generation of controlled atmospheres not only states the basic chemical kinetics and thermodynamics of such important reactions as the water-gas reaction ($CO + H_2 \rightleftharpoons CO + H_2O$) and the producer reaction ($C + CO \rightleftharpoons 2CO$), but also makes a critical review of the relative merits both for laboratory and works of the various alternative types of atmosphere, such as cracked ammonia, hydrocarbon gas and charcoal gas.

The second section deals with the partial or complete removal of particular constituents, especially sulphur, in various forms from raw gases, and with the methods of control and analysis of the gases.

Most attention is given to the third section dealing with industrial applications, and here the author considers the variation in atmospheric control necessary for carbon and alloy steels in order to keep the steel surface unchanged by the atmosphere. A chapter is then devoted to controlled gas-carburizing, and another to the theoretical and practical considerations of the nitriding process. Finally, a brief account is given of the probably less important, or at least less studied, subject of atmospheric control for non-ferrous alloys, including the control for checking the loss of zinc from brass and the avoidance of embrittlement of copper by the action of hydrogen.

This book will prove of great value to all those who are in any way interested in such processes as bright-annealing, bright-hardening, gas-carburizing and nitriding, as well as to metallurgists in general who are seeking to extend the range of application of their materials.

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